BRITISH WELDING JOURNAL

1958 NOVEMBER

VOL. 5 NO. 11

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LOW TEMPERATURE PROPERTIES OF Al-Mg ALLOYS

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J. E. Tomlinson and D. R. Jackson
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R. E. Lismer

Residual Welding Stresses
by V. I. Trufyakov
(Translated from the Russian)

Recommendations for X-ray Examination of Light Alloy Welds

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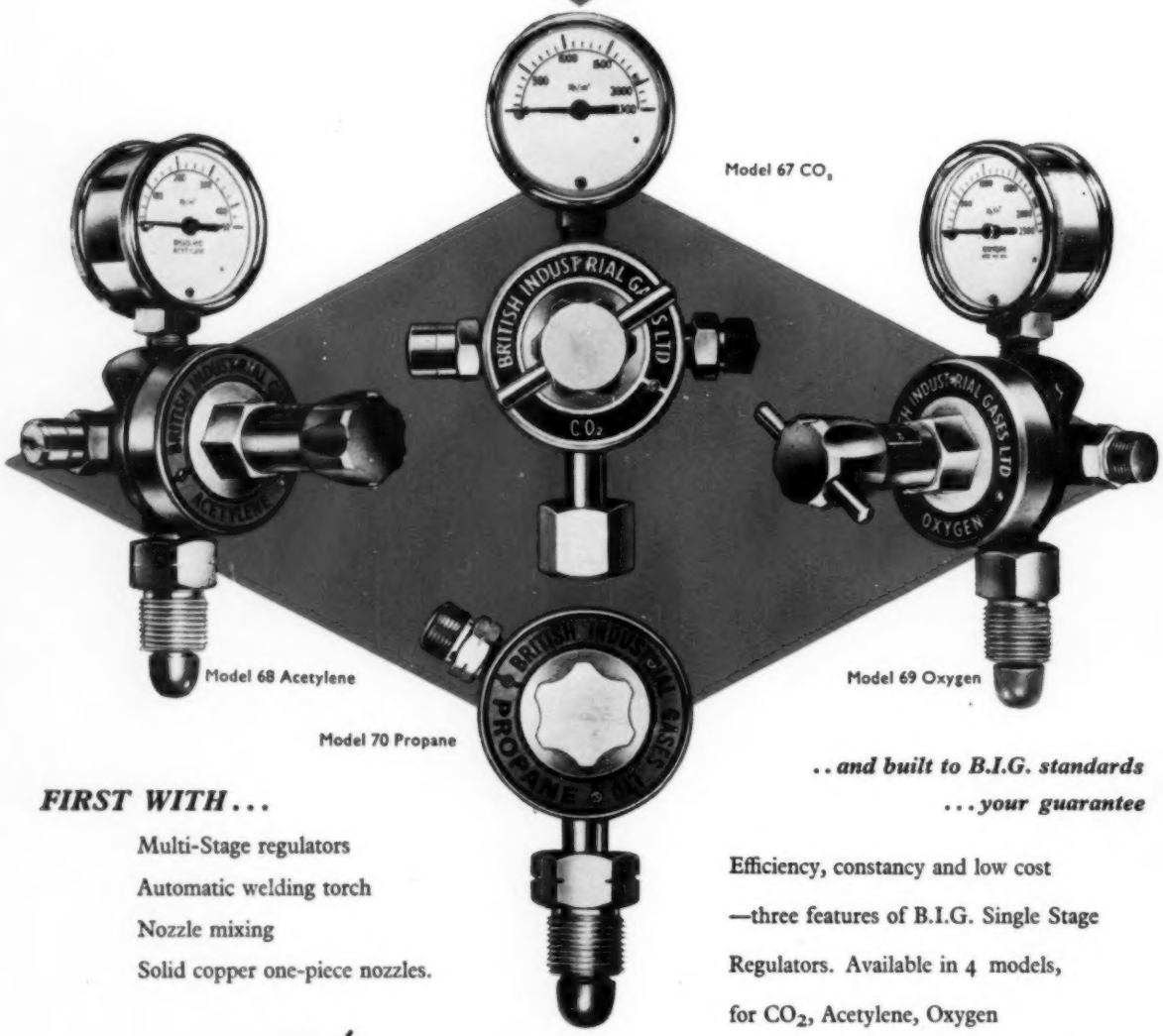

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Inside front cover

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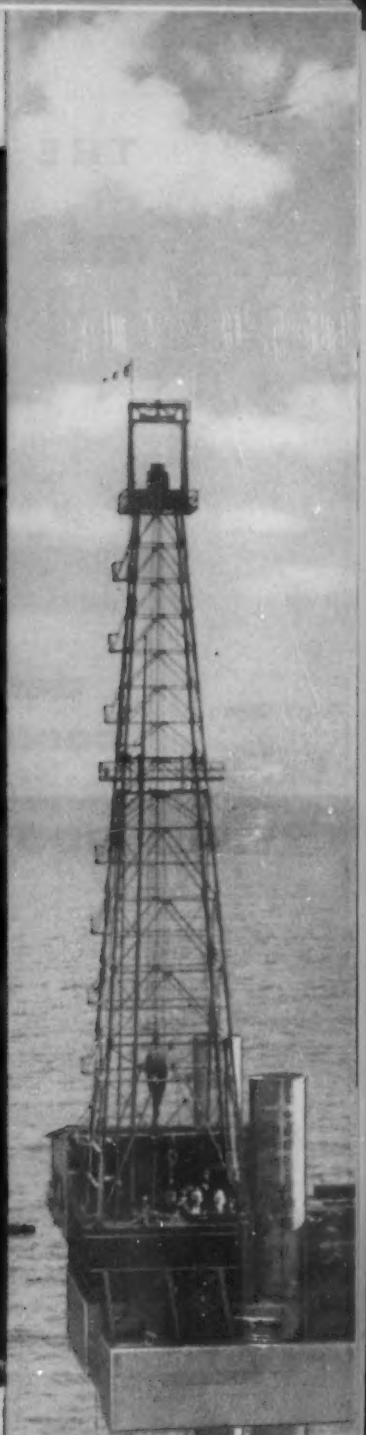

The main photo shows a close-up of a Tubular Steel Production Tower built by the Eonit Engineering Co. Ltd., and welded throughout with Rockweld Electrodes. For use in the Persian Gulf, this tower forms part of a construction which will replace the drilling rig shown above it all is found.

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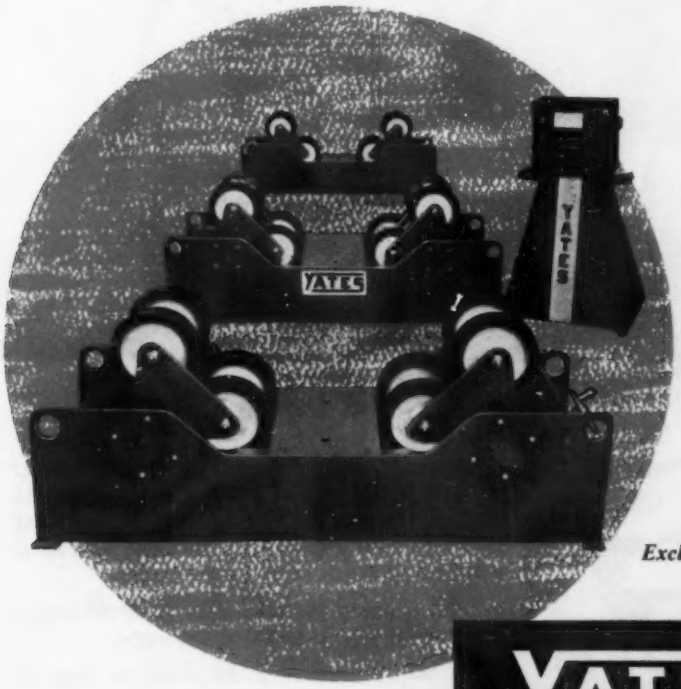

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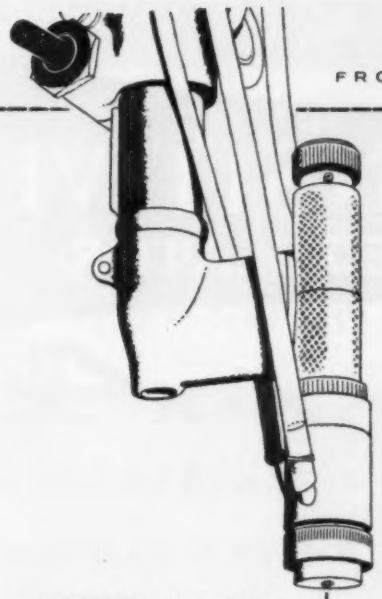

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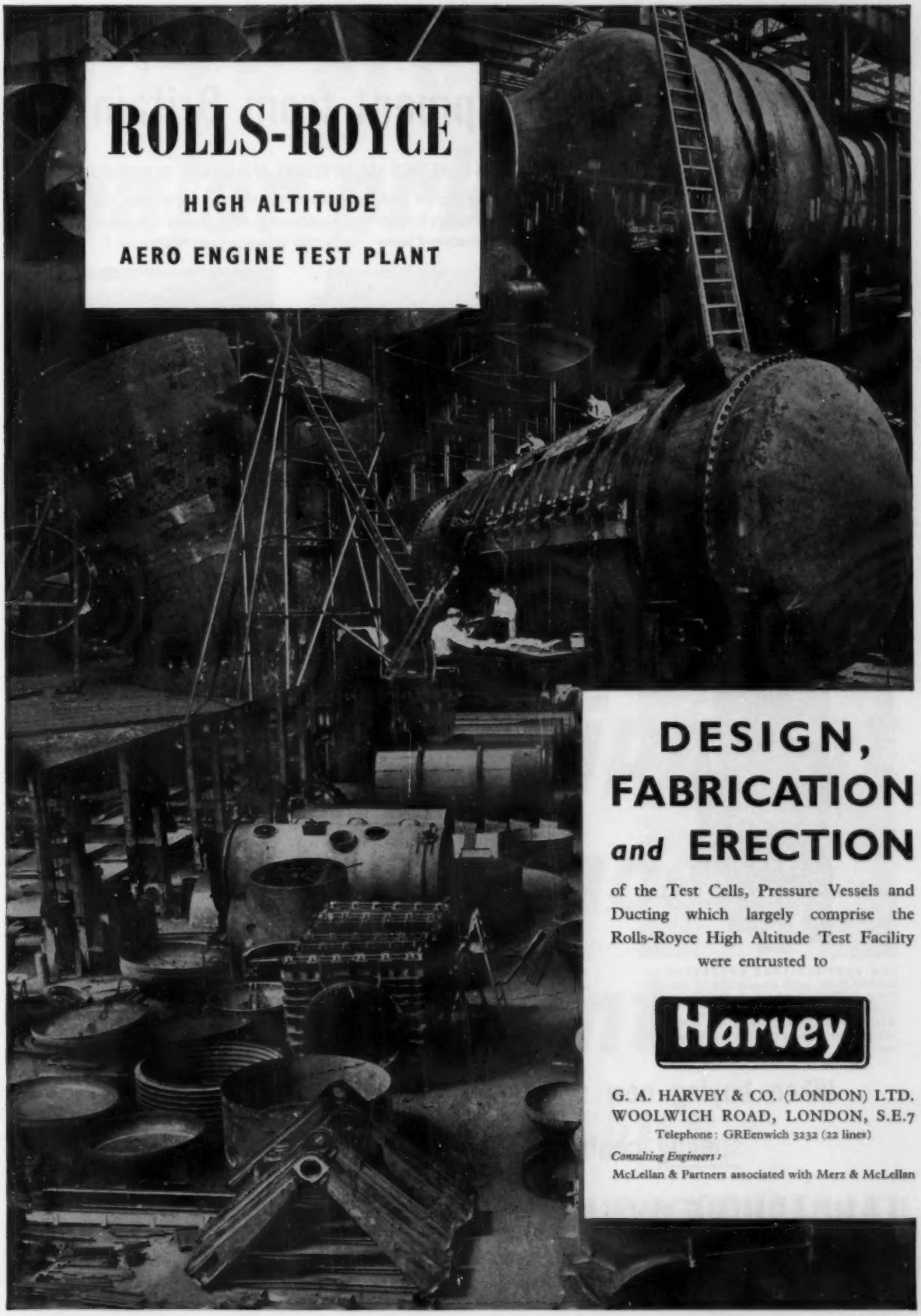

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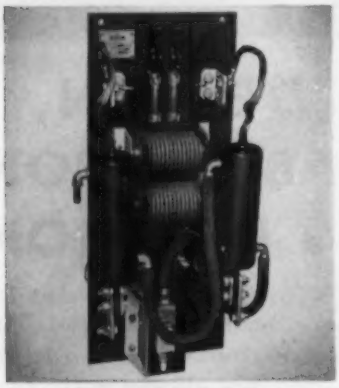

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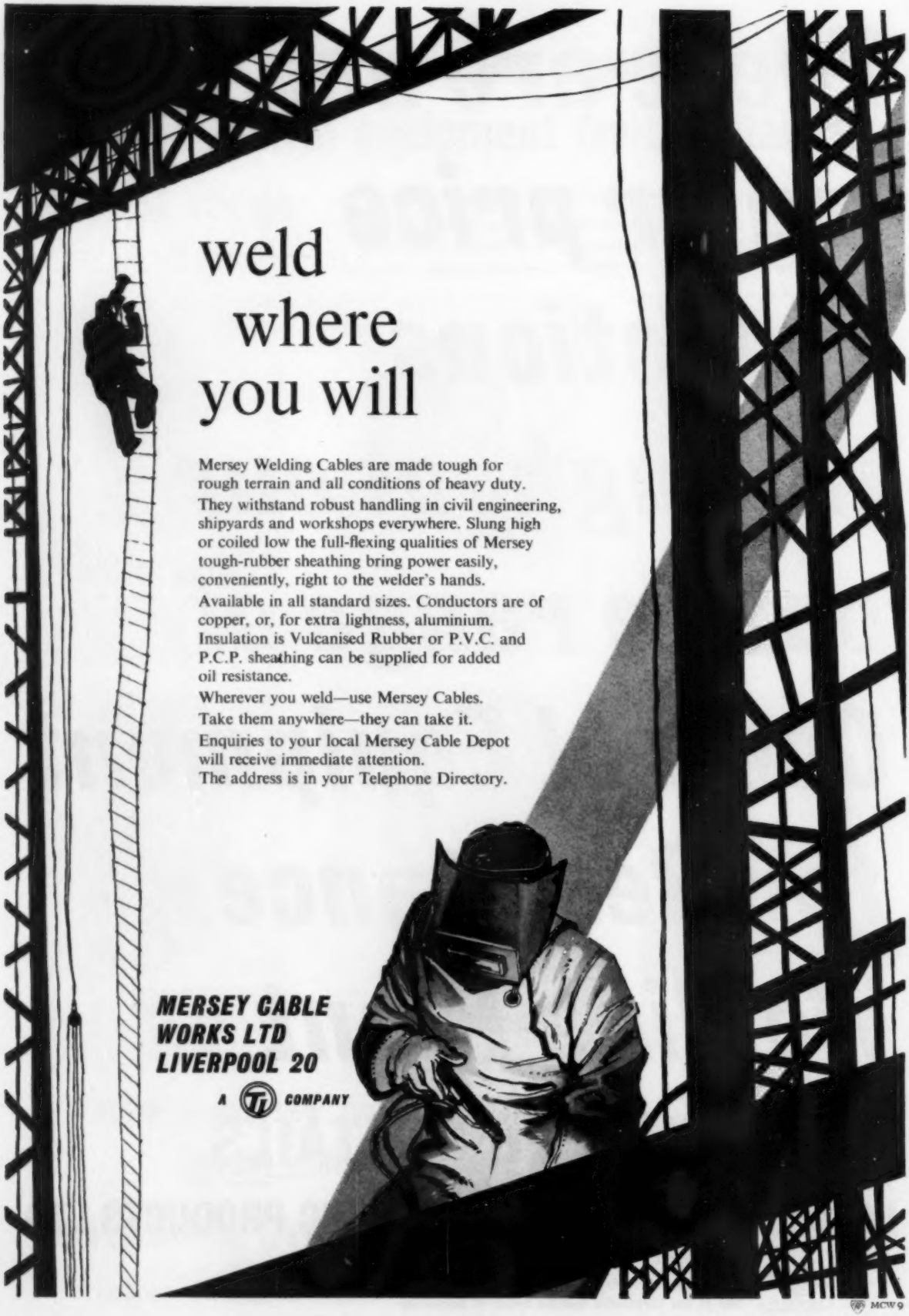

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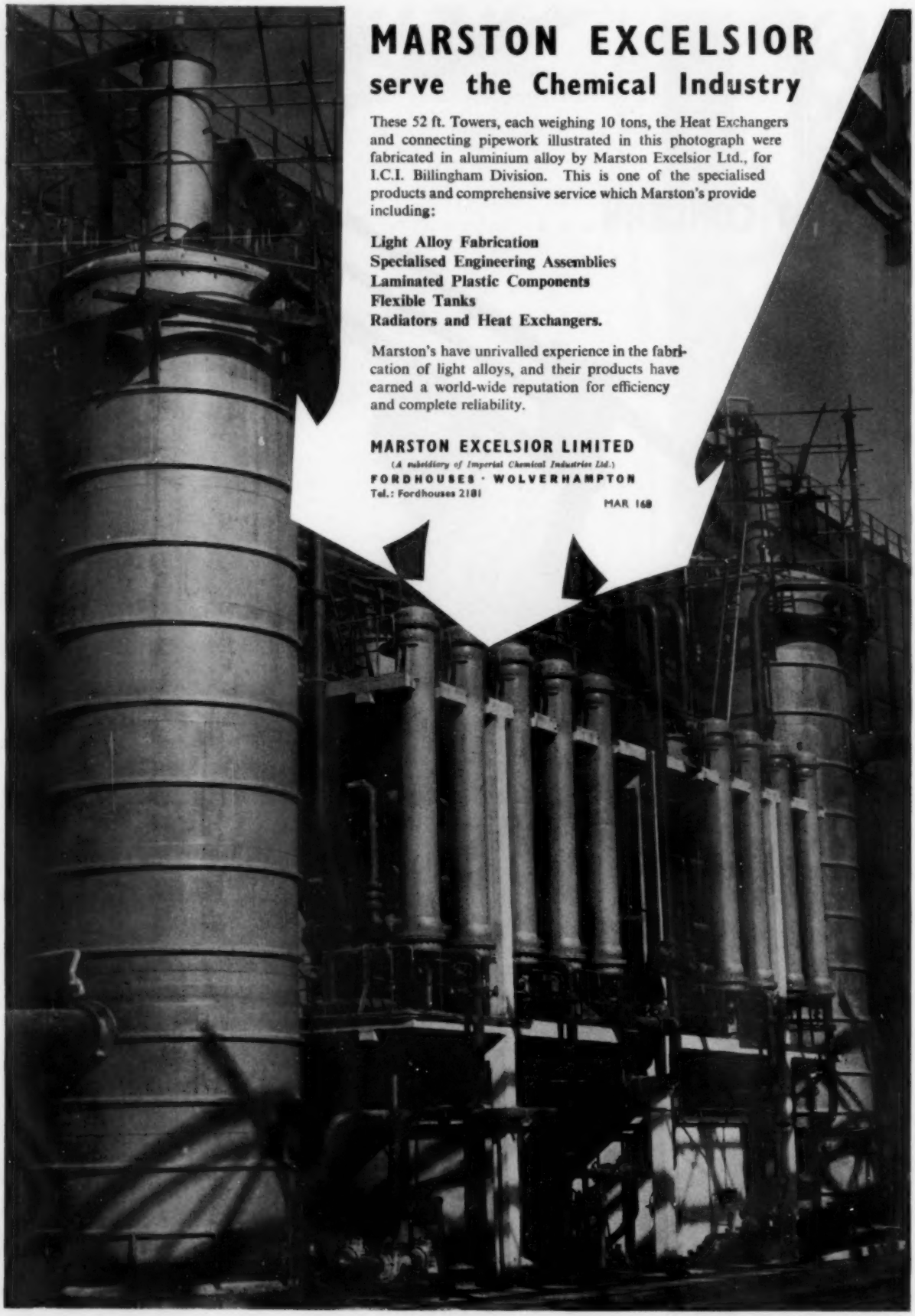

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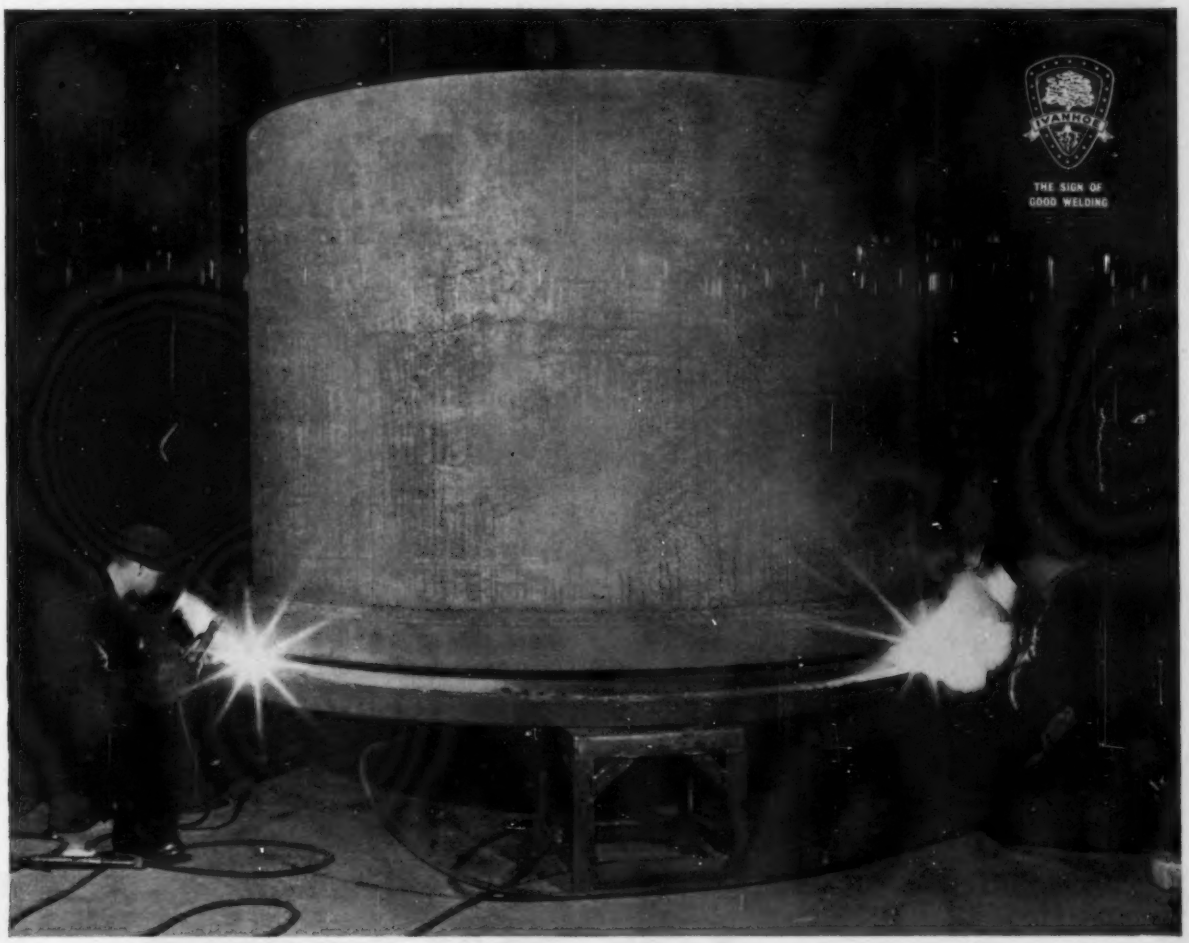

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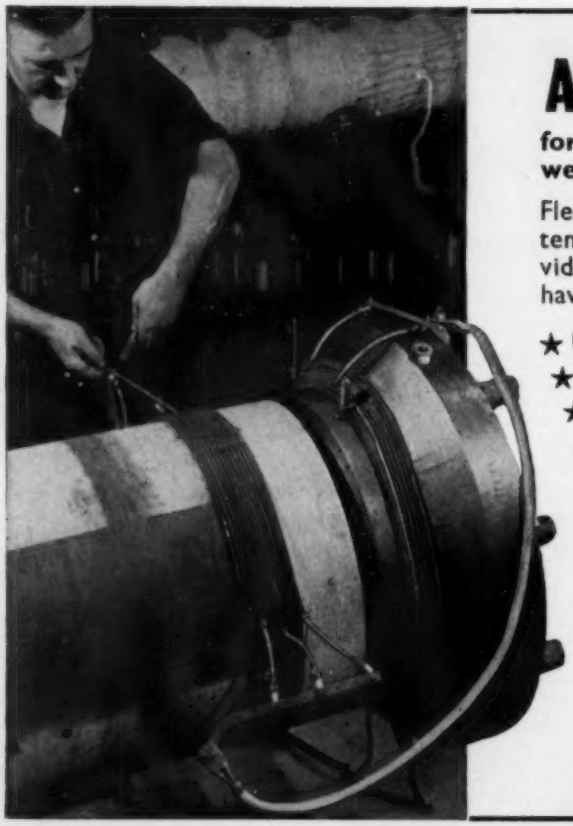

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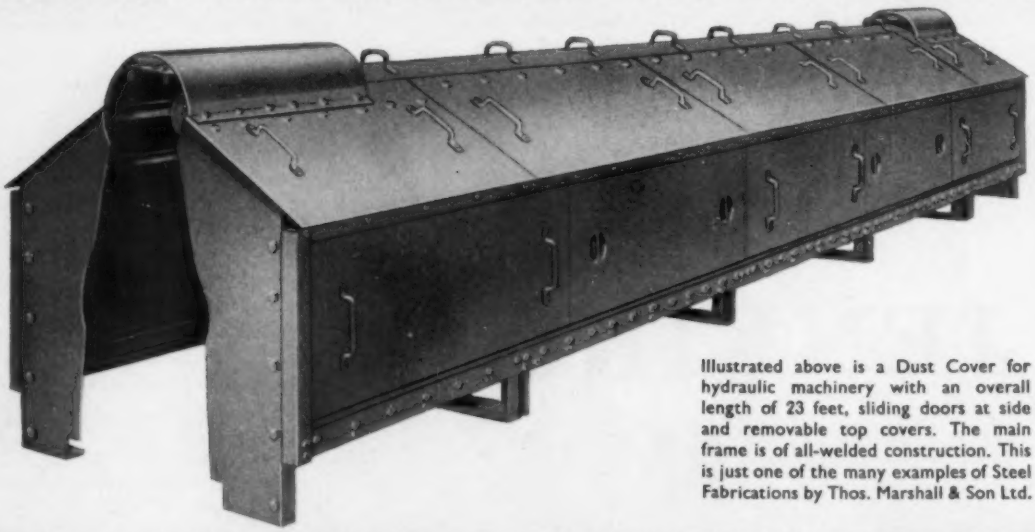

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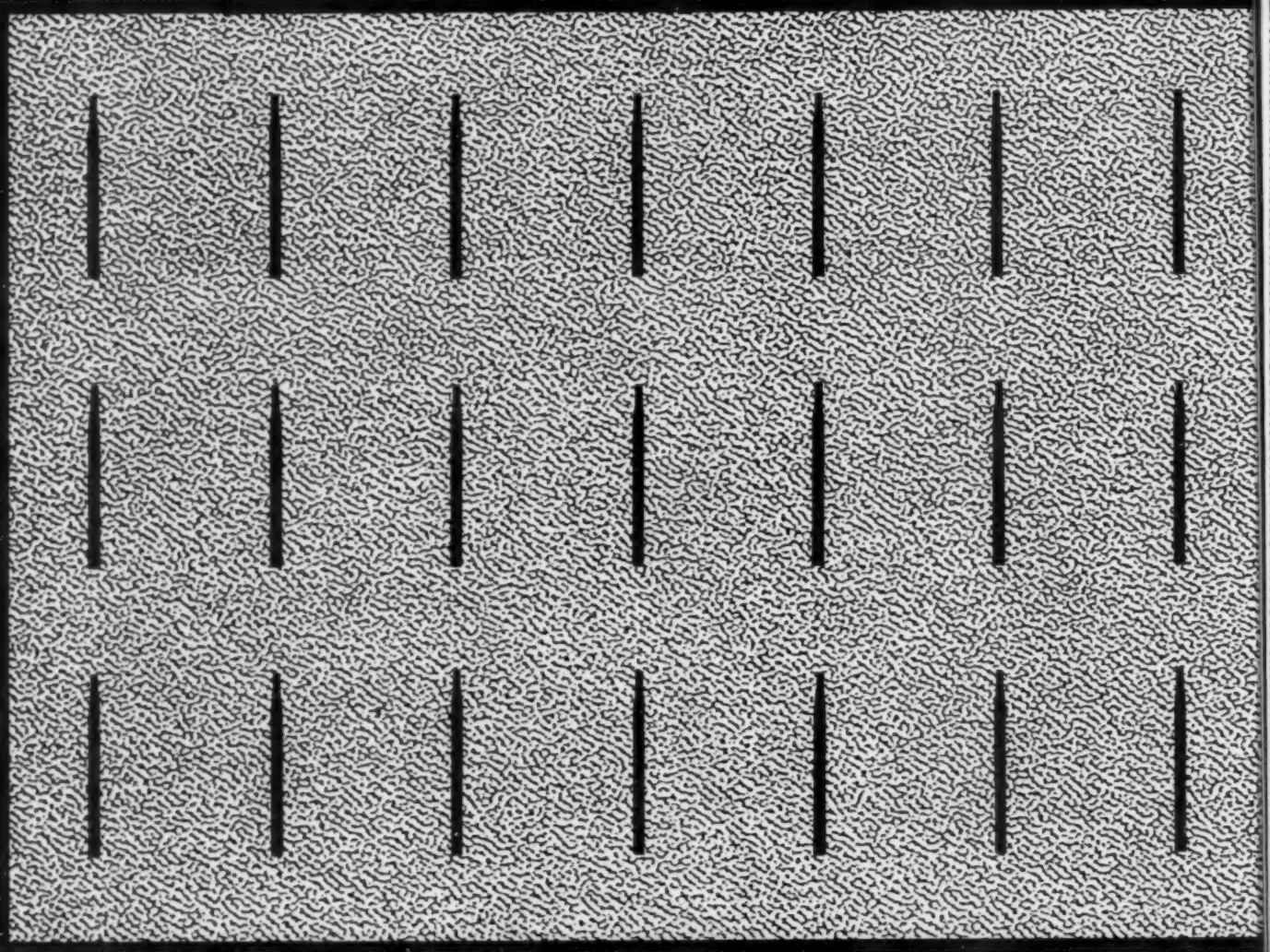

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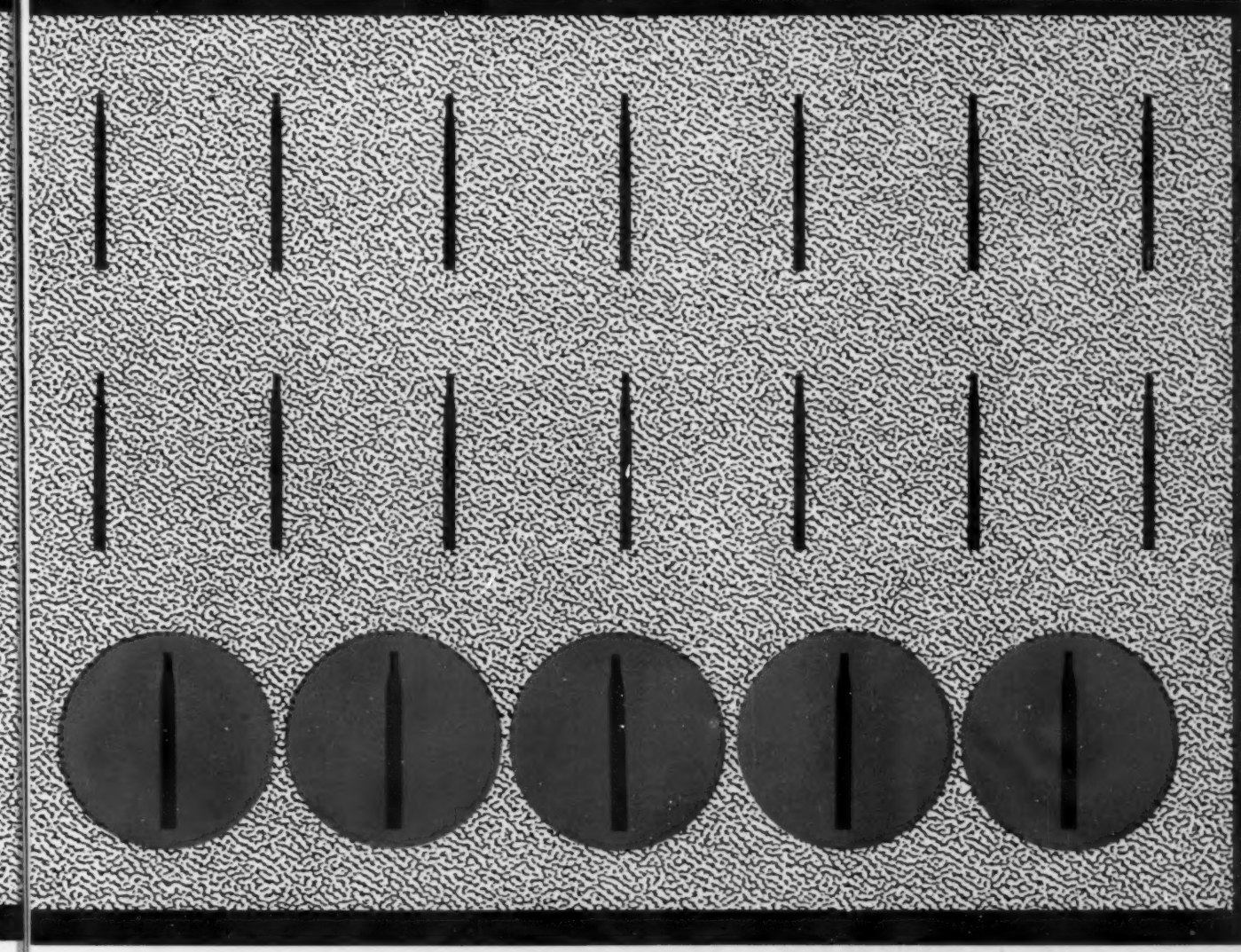

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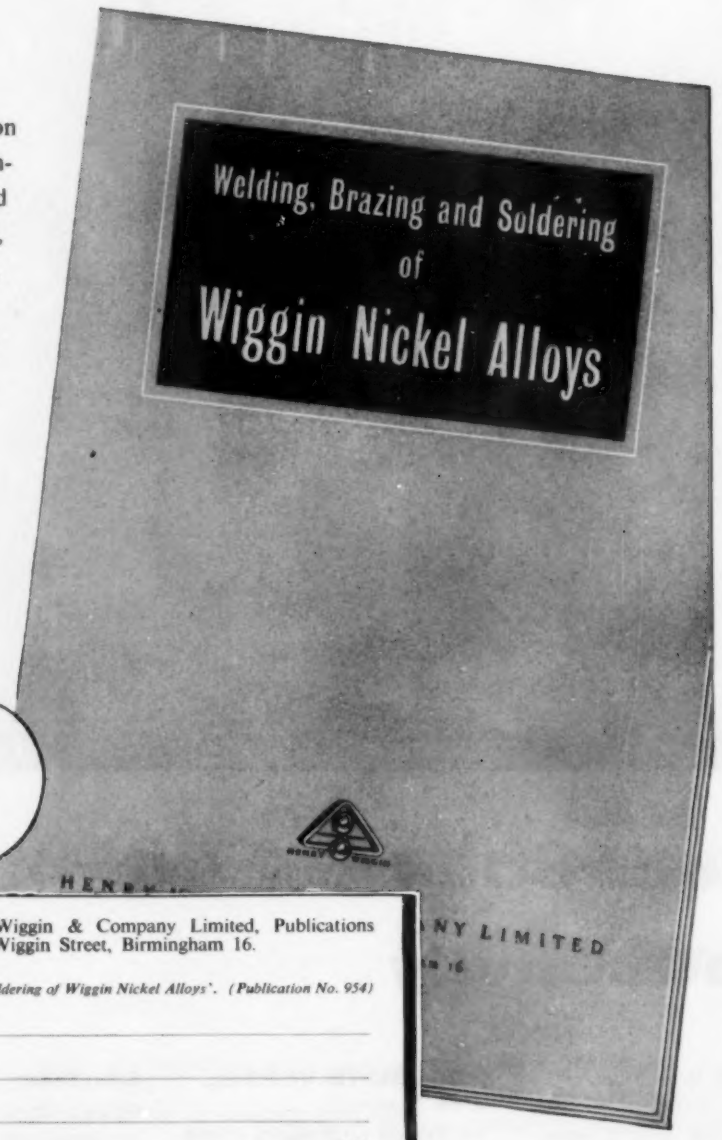

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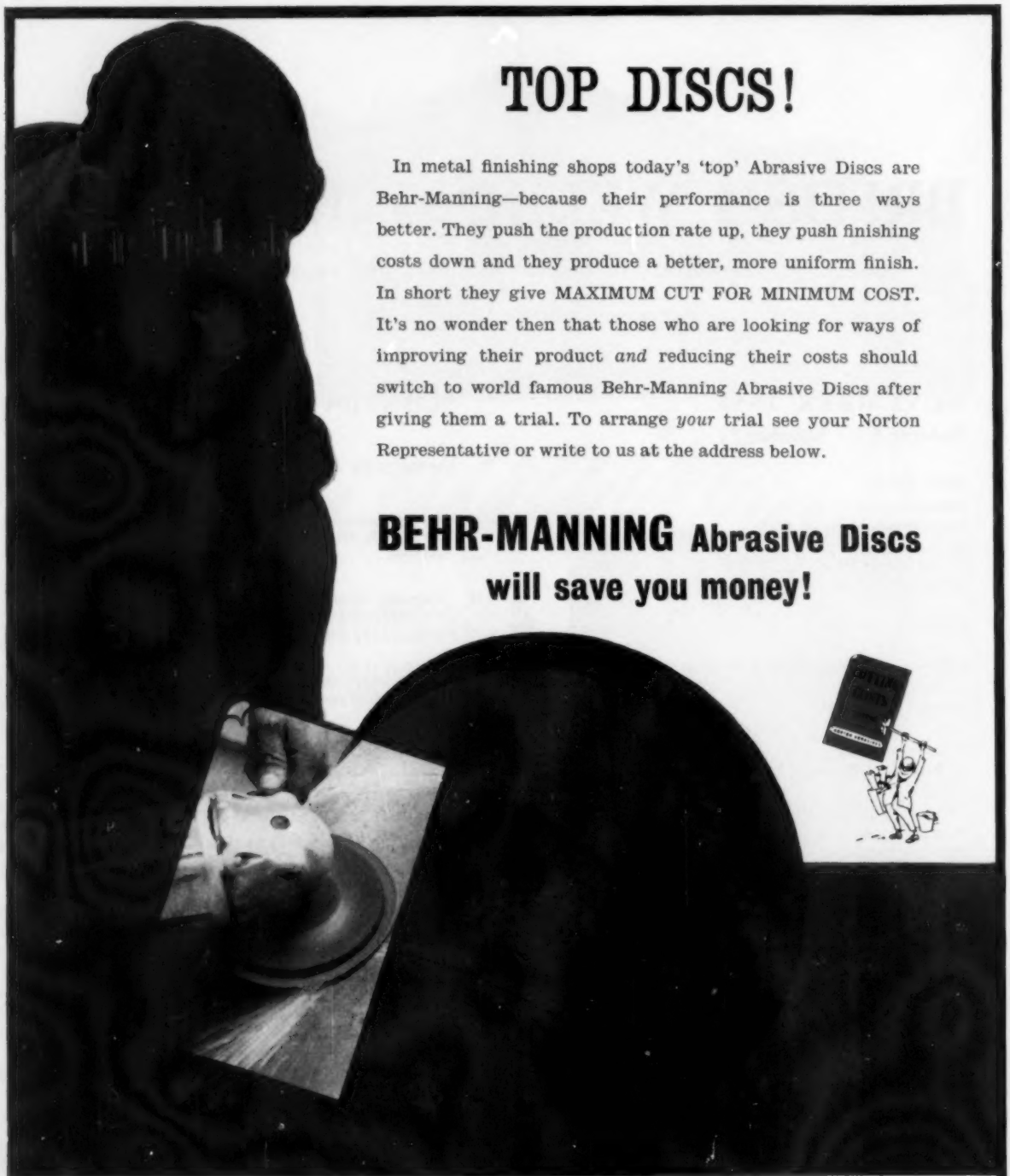

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B.W.R.A. REPORT

Recommendations for the X-ray Examination of Fusion Welded Joints in Light Alloys

SYNOPSIS

RADIOGRAPHIC techniques are in successful commercial use for inspecting fusion welds in light alloys. This B.W.R.A. report gives recommendations for the radiographic inspection of fusion welded butt joints in aluminium and magnesium alloys for thicknesses from $\frac{1}{8}$ in. to 4 in.

Three techniques are described, covering work ranging from the routine to the most critical. The recommendations do not cover gamma radiography because no suitable gamma ray sources are currently available for routine work. The techniques can sometimes be used for inspecting the junctions of pipes or branches with plates or cylinders, but are not applicable to spot, resistance, or stud welds.

Scope

THE following recommendations are based primarily on radiographic techniques which have been, and are being, used successfully in industry for the inspection of fusion welded butt joints in aluminium and magnesium alloys for thicknesses within the range of $\frac{1}{8}$–4 in. Three techniques are described, to cover work ranging from the routine to the most critical; these different standards are achieved by using three different types of non-screen X-ray film whilst keeping all other factors constant.

Non-screen (direct-type) X-ray films are recommended throughout.

The recommendations do not cover gamma-radiographic examinations because no suitable sources are currently available for routine work.

The recommendations cannot be applied to every type of butt welded joint; *e.g.*, it may be impossible to apply them to butt joints in cylinders of small diameter, for the examination of which special techniques are required. They may be applied, in some instances, to the examination of the junctions of pipes or branches with plates or cylinders, but are not applicable to spot, resistance, or stud welds.

Protection

All radiographic work should be carried out in accordance with the current legislation or recommendations appertaining to the protection of operators and other workers from the effects of the X-rays used.

Surface Preparations

The surface finish of the welds will be dependent upon the type of service to which the welds are to be subjected. The provision of a smooth surface, leaving the weld metal flush or with a smoothed reinforcement, facilitates radiographic interpretation, but there are applications where such a refinement is not necessary.

Location of the weld in the radiograph

Markers, usually in the form of lead arrows or other symbols, should be placed alongside, but just clear of, the outer edges of the weld to identify its position.

Identification of the radiograph

Each section of weld radiographed should have suitable symbols affixed to identify:

(*a*) the job or workpiece,
(*b*) the joint,
(*c*) the section of the joint.

Report FE.18RP/66B/58 of the British Welding Research Association, issued to members in March 1958.

The symbols, consisting of lead letters or numerals, should in general be placed on the surface facing the source of radiation.

Marking out

In general, permanent marking by stamping the workpiece will provide reference points for the accurate relocation of the position of each radiograph. Where the nature of the material and its service conditions render stamping undesirable, other suitable means of relocation of the radiographs must be sought.

Penetrameters

A penetrameter of a type agreed between the contracting parties should be placed at one or both ends of every section radiographed, on the surface facing the source of radiation and adjacent to the weld. Only where this surface is inaccessible should the penetrameter be placed on the film side. It must be recognised that in such a position the penetrameter may give a less reliable indication of penetrameter sensitivity. The position of the penetrameter should be noted in the radiographic report.

Type of film

The X-ray film should be of the non-screen type (direct-type). The choice of the varieties of such films available should depend on the technique chosen, as follows:

Technique 1: ultra-fine grain type.
Technique 2: fine-grain type.
Technique 3: medium-grain type.

Very fast non-screen films having a grain coarser than the film intended for technique 3 should not be used.

Technique 1 is intended for the most critical examinations and is particularly valuable for thin sections; technique 2 is suitable for routine inspection, whilst technique 3 is less critical.

Film-holders and cassettes

The front of the film-holder or cassette may be so absorbent as to act as a filter and so impair sensitivity. For this reason, envelope-wrapped films should be used for thin sections. Alternatively, cardboard exposure holders, plastic cassettes or metal cassettes may be used subject to tests showing that the front has no adverse effect on sensitivity for the weld section being examined.

When lead intensifying screens are used, metal cassettes are desirable to ensure good screen film contact; the filtration effect of the front of the cassette will be negligible relative to that of the front lead intensifying screen.

Intensifying screens

In general, intensifying screens are not necessary except for sections of aluminium over 2 in. thick and of magnesium over 3 in. thick, when lead intensifying screens may be used, primarily as filters to minimize the adverse effects of scattered radiation on the image. These screens should be of lead foil approximately 0·004 in. thick for the front screen and 0·006 in. thick for the back. For 100 kV sources or greater, slightly shorter exposures may be achieved by using a back lead intensifying screen, with no front screen, or an especially thin front screen (0·001 in. thick lead or tin foil).

Target–film distances*

For a focal spot with a maximum effective dimension of 4 mm, the target–film distance should be not less than 32 times the throat thickness of the weld, except for throat thicknesses exceeding 2 in. when a minimum target–film distance of 64 in. may be used. Distances for other focal spots should be in direct proportion to their maximum effective dimensions.

Object–film distance

The distance between the film and the adjacent weld surface should be as small as possible. Where a gap between the surface of the weld and the cassette is unavoidable, the minimum target–film distance should be increased in the ratio of:

$$\frac{\text{throat thickness} + \text{gap}}{\text{throat thickness}}$$

Alignment of X-ray beam

The beam of radiation should be directed to the middle of the section under examination and normal to the plate surface at that point, except in special examination for certain defects which it is known would be best revealed by a different alignment of the beam; *e.g.*, defects at a fusion face, when the exposure may be made with the beam directed along the fusion face.

Overlap of film

The separate radiographs of a welded seam should overlap sufficiently to ensure that no portion of a seam remains unexamined.

Length of weld examined

The length of weld which may be examined at one exposure should be such that the thickness of the metal at the extremities of the exposed area, measured in the direction of the incident beam at that point, should not exceed, by more than 6%, the actual thickness at that point.

Interception of unwanted and of scattered radiation

The film should be shielded from all back-scattered radiation by an adequate thickness of lead of at least

* The distances quoted are for a geometric penumbra of 0·005 in. to allow for examination of the image by magnification.

Table I

Relationship between material thickness and kV

Thickness, in.	*Technique 1 (Ultra-fine grain film), kV*	*Technique 2 (Fine-grain film) kV*	*Technique 3 (Medium-grain film), kV*
up to ¼	70	60	50
¼–½	80	70	55
½–¾	95	80	60
¾–1	105	90	65
1–1¼	115	100	70
1¼–1½	125	110	80
1½–2	145	130	90
2–2½	165	150	110
2½–3	185	170	120
3–4	230	210	150

0·1 in. When masking is used to limit the irradiated area of the specimen and so reduce the effect of scattered radiation internally, it should be so arranged as not to cause any obscuring of radiographic detail.

Density of radiograph

Exposure conditions should be such that the density of the image of the sound weld metal is not less than 2·0 exclusive of fog density and inclusive of base density.

The fog density of the processed unexposed film should not exceed 0·2 inclusive of base density.

Tube voltage

The maximum tube voltage for pulsating X-ray generation circuits should not exceed the values given in Table I, and should be chosen so that the exposure is not less than 15 mA–min. For fully rectified circuits, the maximum voltages quoted should be reduced by 10%. For magnesium alloys, generally lower kilovoltage values will apply.

Processing

Films should be processed in accordance with recognised good practice using a standard type of X-ray film developer. The processing solutions should be maintained in an efficient working condition and either the film or the processing solution should be agitated during development. The development times should be not less than 5 and not more than 8 min at 68°F. (20°C.) or the equivalent at other temperatures. These times may have to be reduced when using higher energy developers, depending on the type of film and type of developer employed.

The radiographs should be free from imperfections arising from handling and processing or other defects which would interfere with interpretation.

Viewing

The radiographs should be examined by diffused light in a darkened room and the illuminated area should be masked to the minimum required for viewing the radiographic image.

The brightness of the illuminated radiographs should be not less than 1 candela/sq.ft (10·76 candela/sq.m) and preferably greater than 10 candela/sq.ft (107·6 candela/sq.m).

B.W.R.A. REPORT

Welded Joints and Residual Stresses

By V. I. Trufyakov

RUSSIAN articles on welding have a freshness of outlook which has led the B.W.R.A. to arrange for the translation of extracts from several sources. If space allows, further translations will be published from time to time. The article below, which appeared in "Automatic Welding" (*Avtomaticheskaya Svarka*) 1956, No. 5 (October), describes the effects of residual stresses on arc-welded joints of various sizes, the largest being a butt joint in a 12 in. deep joist.

THERE is a widespread opinion that the strength of welded joints in a given steel is determined basically by the magnitude of the stress concentrations created by the joint shape and by various weld defects such as undercutting, pores, and cracks. By comparison, the influence of other factors, including residual stresses, is considered to be secondary. Because of this, and the low power of the testing machines in use, the fatigue strength of welded joints and even of full-size structures is normally determined from experiments on small test pieces.

The object of this paper is to show that the strength of a welded joint is determined largely by the magnitude and nature of the residual stresses, for these cannot be disregarded when evaluating the fatigue strength of joints, and must be taken into account in the experimental determination of the strength limits.

EFFECT OF RESIDUAL STRESS AT A WELD NOTCH

The research carried out by Kudryavtsev[1] has proved that residual stresses have a conspicuous influence on fatigue strength in a test piece with a stress raiser. He came to this conclusion on the basis of results from small test pieces in which the residual stresses were mainly created by high-frequency induction heating, speedy cooling, shot-blasting and other methods.

It has also been established[2] that where residual stresses result from welding, the fatigue strength drops if the weld is notched.

It may be assumed that for each severity of notching there exists one specific residual stress effect. In actual welded joints the severity of notching is determined by

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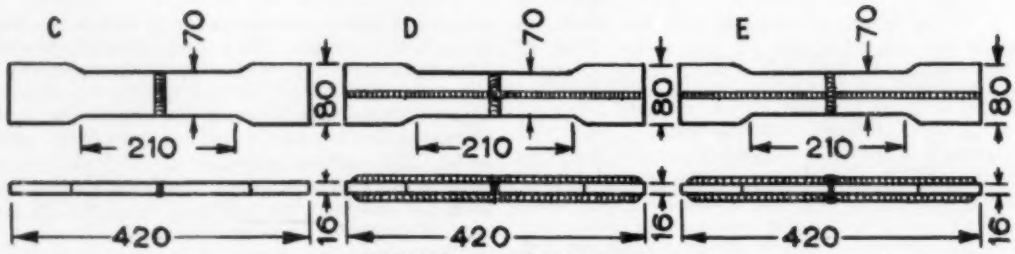
The author is at the Arc Welding Institute of the E. O. Paton Order of the Red Banner of Labour; Academy of Sciences of the Ukrainian SSR.

Translated by Major J. H. Dixon for the British Welding Research Association.



1—Test pieces of series C, D, and E

the joint shape. Two groups of experiments were therefore made to verify that the residual stresses were effective when the stress raiser was created by the joint shape and not by a sharp artificial notch.

First group of tests

In the first group three series of butt-welded test pieces *C*, *D*, and *E* (Fig. 1) made from 16 mm thick type M16S steel were used. The chemical composition and mechanical properties of the material are given in Tables I and II.

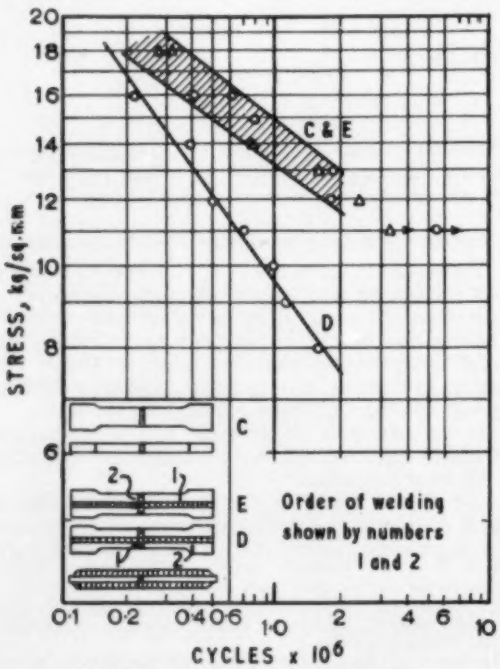
Each test piece in series *C* was a normal butt joint automatically welded under the following conditions: Current, 650–700 amp; arc voltage, 32–34 V; welding speed, 13 in./min. The junction between the parent metal and the weld metal reinforcement is a natural stress raiser in this type of joint.

Series *C* consisted of 12 test pieces. Six of these were cut by milling cutter from the same plate, which had a central butt weld. The other six specimens were prepared separately.

In all these test pieces, the residual stresses transverse to the weld were low. Measurements made by the X-ray method proved that the highest transverse residual tensile stresses in the samples cut from the same plate did not exceed 1·6 tons/sq.in., but they were somewhat higher in the separately prepared test pieces. It is assumed as a first approximation that the test pieces in series *C* had no residual stresses in the longitudinal direction.

To create large residual stresses, additional longitudinal beads were deposited on the test pieces in series *D* after the butt welds had cooled. The welding conditions were: Current 620–650 amp; arc voltage, 32–34 V; welding speed, 22 in./min. In this instance the residual longitudinal stresses approached the yield point of the parent metal.

Test pieces in series *E* differed from those in series *D* in the order in which the welding was performed. The beads were first deposited on the plate, which was



2—Results of tests on series C, D, and E

Table I

Composition of parent material

Plate Thickness, mm	*Analysis, %*				
	C	*Mn*	*Si*	*S*	*P*
16	0·18	0·53	0·26	0·031	0·021
26	0·16	0·49	0·25	0·033	0·024
46	0·19	0·44	0·15	0·048	0·013

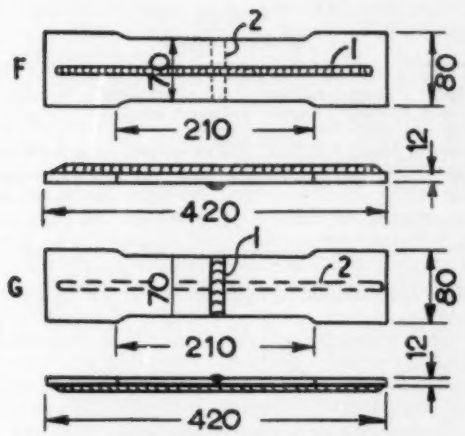
Table II

Mechanical properties of parent material

Plate Thickness, mm	*Yield Point, tons/sq.in.*	*U.T.S., tons/sq.in.*	*Elong., %*	*Red. in area, %*	*Impact strength at 20°C. Kg.m/sq.cm*
16	15·7	27·6	38·5	64·0	10·0
26	15·8	26·4	34·4	64·0	11·8
46	14·8	26·0	42·2	66·8	8·1

then cut into two pieces which in turn were butt welded together (Fig. 1). In this way the geometry of the test pieces was the same as in series *D*, but the longitudinal residual stresses were the same as those in the test pieces containing butt welds only.

The test pieces in all three series were tested in alternating stress cycles in a machine made to N. N. Afanasev's design. The results are given in Fig. 2, and show that the test pieces containing butt welds only (series *C*) did not give a clearly defined S–N curve. If lines are drawn to enclose the extreme points, a zone covering the scatter of results for series *C* is obtained. All the points relating to the test pieces in series *E* also fall in this region. The concentration of stresses in the



3—Test pieces of series F and G

intersecting welds is therefore the same as that in the normal butt welds.

The fatigue strengths* of test pieces in series *C* and *E* (which have low residual stresses) are identical and are somewhere between 7·3 and 8·1 tons/sq.in.

The test pieces in series *D* give different results. The S–N curve diverges increasingly from the scatter zone as the magnitude of the alternating stresses drops. The fatigue strength drops to 4·8 tons/sq.in. at 2×10^6 cycles. This sharp reduction in the fatigue strength of butt joints can only be ascribed to the residual stresses.

These results are confirmed by the appearance of the test-piece fractures. In specimens without residual stresses the fatigue crack is almost uniformly distributed across the width; in test pieces from series *D*, however, fatigue cracks formed in the areas of highest residual tensile stress, under the notches formed by the edges of the weld beads.

Second group of tests

The test pieces in this group (series *F* and *G*, Fig. 3) were also prepared from type M16S steel. Each test piece had two beads; one transverse, the other longitudinal, deposited on alternate sides of the plate. For series *F* the longitudinal bead was deposited first, and the transverse bead was deposited on the cooled metal. The reverse order of welding was adopted for series *G*.

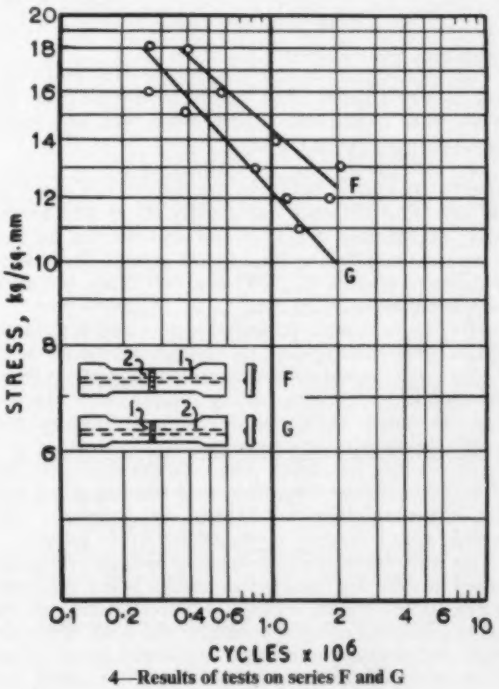
Since higher residual stresses are set up in the transverse weld zone a reduced fatigue strength in the series *G* test pieces would be expected, and experiments have confirmed this. The results of the tests for both series are given in Fig. 4.

The fatigue strength of test pieces from series *F* is reduced to 7·8 tons/sq.in., and that for test pieces in series *G* is 6·4 tons/sq.in. Since the different sequence of welding changes only the magnitude of the residual stresses, this reduction in endurance limit can be due to residual stress alone.

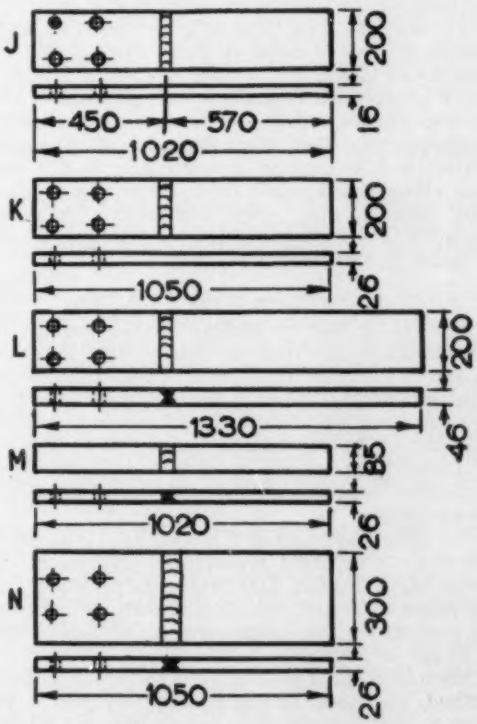
A comparison of the results from test pieces in both groups shows an overall identity. As the magnitude of the alternating stresses drops, the residual stresses intensify their effect and considerably lower the fatigue strength.

Where the severity of notching was the same as that created by the normal reinforcement of a butt weld,

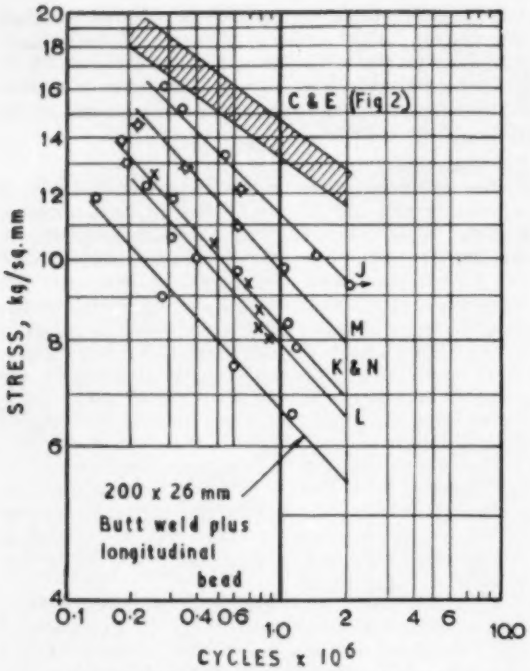
* In this article the term "fatigue strength" refers to the stress value corresponding to 2×10^6 cycles, except where otherwise noted.



4—Results of tests on series F and G



5—Test pieces of series J, K, L, M, and N



6—Results of tests on series J, K, L, M, and N

the residual stresses reduced the fatigue strength to 40%, the magnitude of the residual stresses being of the order of the yield point of the parent metal.

It is evidently impossible to ignore the effect also of smaller residual stresses. In particular, the considerable scatter of the results from test pieces in series *C* and *E* is explained basically by the presence of small residual stresses, differing in the various specimens. It is characteristic that where the value of the residual stresses is fully realized (for example, in test pieces with a longitudinal bead), there is only a small scatter. This indicates that, when determining the fatigue strengths of welded joints, the residual stresses must be taken into account.

SIZE EFFECT

It is known that the residual stresses increase to a definite magnitude with increase in test piece cross section. It was established, for example, in De Garmo's experiments[3] that when a butt weld is pulled, the transverse residual stresses cease increasing only after the width of the test piece has reached 8 in., where the thickness of the plate is 1 in. It is necessary in this connection to investigate the minimum dimensions of the test pieces.

The fatigue strength of a welded joint is at present ordinarily determined from small test pieces. The Tsniis Metal Bridges Laboratory, in particular, uses test pieces of 20 × 80 mm cross-section as the basis of calculations for the fatigue strength of welded joints in bridge structures.[4]

When calculating welded joints by the limiting state method, the rated fatigue strength adopted is very close to the corresponding values obtained likewise from small test pieces.

Table III
Welding conditions

Plate Thickness, mm	*Edge Preparation*	*Current, amp*	*Arc voltage, V*	*Welding Speed, in./min.*	*Number of Passes*
16	None	750	32–34	15	One each side
26	Vee	800	34–36	13	Six and one backing pass
46	Double V	800	34–36	13	Five each side

Since the residual stresses do not nearly reach their maximum in such test pieces, it is to be expected that the fatigue strength will drop as the cross-section is increased.

Experimental data [5, 6] show that the fatigue strength of a given welded joint varies depending on whether it is in a large joist or in a small test piece. Normally the joists withstand a considerably smaller number of cycles before failure than the test pieces. It is suggested that the lower fatigue strength of the welded joints in actual joists indicates the importance of the size factor.

In view of the foregoing, two problems must be solved:

(i) Does the fatigue strength of a welded joint really drop considerably as the cross-section of the test piece increases? If so, what are the limiting dimensions of the test piece beyond which no further reduction occurs in fatigue strength?

(ii) What part do the residual stresses play in the lowering of fatigue strength with increase in the cross-section of test pieces?

Five series of specimens were tested to investigate the first question (Fig. 5). They were prepared from type M16S (GOST 6713–53) steel, whose chemical composition and mechanical properties are detailed in Tables I and II. Table III gives the welding conditions for the butt welds which were deposited by the automatic submerged-arc process.

The weld reinforcement was not removed and was a natural stress raiser. The nature of the junction between weld metal and parent metal was practically identical in all the test pieces.

The test pieces can be divided into two groups: series *J*, *K* and *L* being in the first. They were all 200 mm wide and were respectively 16, 26 and 46 mm thick. Test pieces *M* and *N* and also *K* from the first group are included in the second group. These were of different widths (85, 200 and 300 mm), but their thicknesses were all 26 mm.

All specimens were subjected to reversed bending in almost symmetrical cycles, the coefficient of asymmetry of the cycle being $-0{\cdot}950 \pm 0{\cdot}020$. The tests were carried out on a fatigue-testing machine with the test piece mounted as a cantilever, and loading was applied by an electro-magnet.

The loading frequency was synchronized with the natural frequency of oscillation of the test piece, and was in general between 33 and 44 cycles/sec. The stresses were measured by resistance strain gauges.

The tests were stopped when a fatigue crack appeared in the test piece; the cracks being disclosed during testing by an ultrasonic defectoscope. All the cracks appeared along the edge of the weld reinforcement, and always commenced in definite zones; either at the centre of the test piece or at both the centre and

the edges. The fatigue tests were continued to 2×10^6 cycles, and the results from test pieces of all five series are given in Fig. 6.

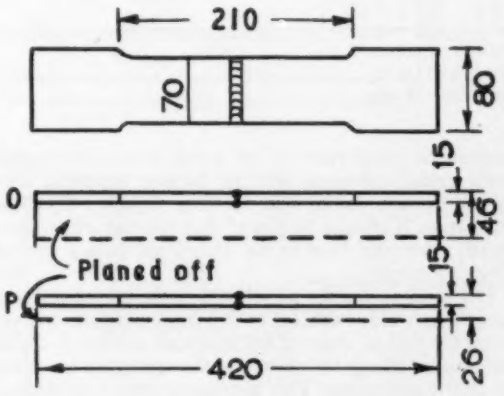
Each series has its own definite S–N curve, and it is evident that the fatigue strength drops considerably with increase in the dimensions of the test piece, whether by widening or by thickening. A butt-joint fatigue strength of 7·6 tons/sq.in. is obtained for small test pieces (the zone of scatter of results from small test pieces of cross-section 70×16 mm has been included at the top of Fig. 6 for comparison); but in large test pieces it drops to almost 3·8 tons/sq.in. Test pieces of cross-section 200×16 and 85×26 mm provide intermediate values.

The drop in fatigue strength ceases above specific test-piece dimensions. Almost identical fatigue strengths were obtained from test pieces of 200×26, 300×26, and 200×46 mm size. Therefore test pieces not less than 200 mm wide and not less than 26 mm thick must be used for determination of the fatigue strength of a butt joint, for all smaller cross-sections give higher values. Increase in cross-section beyond 200×26 does not lead to marked reduction in fatigue strength.

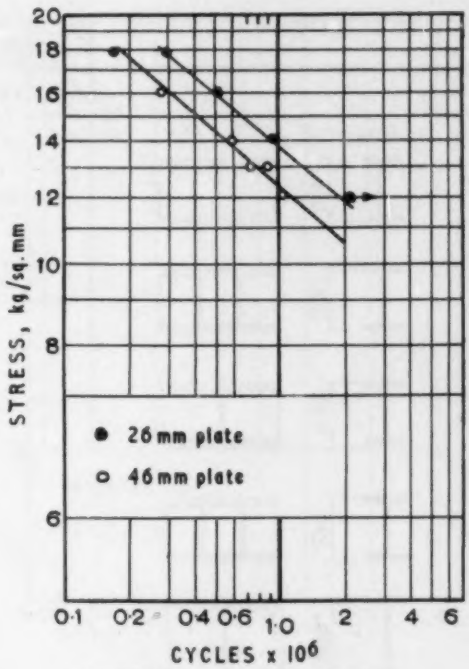
At present a fatigue strength of 7·6–11·5 tons/sq.in. is the starting point for calculating the strength of butt joints.[7] These values were obtained from small test pieces. It is easy to see that test pieces of cross-section 200×26 mm and over will withstand only about one tenth as many cycles; their fatigue strengths are lower by half. It follows from this that the results from small test pieces cannot be used to estimate the fatigue strength of an actual structure.

On the other hand, when comparing these results with those from tests on butt joints in welded joists, it appears that joints in flat test pieces 200×26 mm can have a fatigue strength the same as, or lower than, that of the joists. For example, in Asnis's article[5] welded joists with a transverse bead on the flanges gave a reduced strength of 5·2 tons/sq.in. at 1,000,000 cycles, yet during the present tests a fatigue strength of 5·3 tons/sq.in. was obtained in flat test pieces at 1,000,000 cycles.

The results of tests made on butt-welded joists with various joint designs, taken from a number of projects carried out at the Arc Welding Institute, are given in Table IV. It seems that the joists do not have a shorter life than the flat test pieces and even the reverse is sometimes observed. In many instances therefore it is possible to obtain correct endurance limits for welded joints from comparatively small test pieces. This is fortunate, since costly tests on the joists are not needed and their failure often arises from various accidental causes.



7—Test pieces of series O and P



8—Results of tests on series O and P

RESIDUAL STRESSES, SIZE EFFECT AND FATIGUE STRENGTH

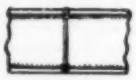
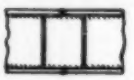
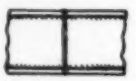
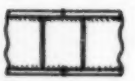
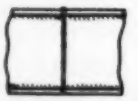
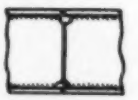
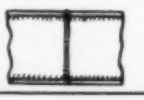
The results from test pieces in the preceding group have shown that the fatigue strength of welded joints drops as the dimensions of test pieces rise to a definite magnitude. Why is this?

Examination of all the factors which may in varying degrees reduce the fatigue strength of test pieces as their cross-section is increased will not be considered. A task like this, which requires clarification of the nature of the size factor, is beyond the scope of the present investigation, which is confined to the part played by residual stresses. It is suggested that these stresses are one of the decisive factors which reduce the fatigue strength of the welded joint as the test piece cross-section is increased.

Since the test pieces used in these experiments were made of metal of different thicknesses, and had different structures and mechanical properties, it is first necessary to clarify the effect of these factors on fatigue strength.

Two series of test pieces containing butt welds (series *O* and *P*) were tested. Specimens in series *O* were made of 46 mm thick steel, planed down to a thickness of 15 mm, with one of the sides left in the natural post-rolling state. This particular part of the

Table IV
Tests on butt welded joists

Serial No.	*Joist Dimensions**	*Elevation of butt Joint*	*Extreme Fibre Stress, tons/sq.in.*	*Cycles to Failure* $\times 10^3$	*Points of initiation of fatigue cracks*
1			±9·0	138	1. At beginning of vertical weld of the web where it is cut. 2. In the flange at the point where the flange weld is broken.
2			±9·0	422	At the upper flange butt weld.
3			±7·9	428	1. In the vertical web weld where it is cut. 2. In the flange where the flange weld is broken.
4			±7·9	570	At the intersection of web and flange.
5			±7·6	500	Along the weld near the flange and web butt welds.
6			±7·6	687	Along the weld near the beginning of the web and the flange.
7			±7·6	320	Along the flange butt weld. Non-fusion and slag inclusions visible in the fracture.
8			±7·2	642	Along the flange butt weld.
9			±7·2	650	Along the flange butt weld.

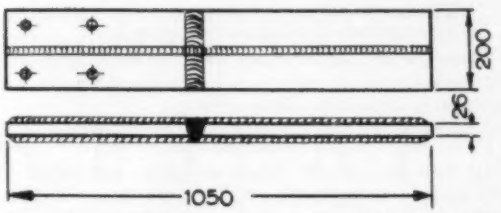
* Specimens 1 to 4 have flanges 160 × 15 mm.
Specimens 5, 6, 7 have flanges 180 × 14 mm.
Specimens 8 and 9 have flanges 125 × 16 mm.

metal, and not the centre of the plate was used, because the fatigue cracks always commence at the surface of the test pieces, where the stress raiser is also located.

Test pieces in series *P* were prepared in exactly the same way, but from 26 mm thick plate. The mechanical properties of the steels and their chemical composition are given in Tables I and II. Figure 7 illustrates test pieces of both series.

The specimens were tested under alternating loading in a machine of N. N. Afanasev's design. The results, given in Fig. 8, show that changes in the structure and mechanical properties of the metal have a comparatively small influence on the fatigue strength. The values deviate from one another roughly by 0·64 tons/sq.in. It should be noted that the test pieces prepared from the thicker 46 mm steel had a higher fatigue strength.

The fatigue cracks in all the samples containing transverse welds (series *J* to *N*) originated, as before, along the line of stress concentrations always in specified zones—along the centre of the test piece or along the centre and edges. This obviously cannot be treated as a chance occurrence, and is possibly due to the fact

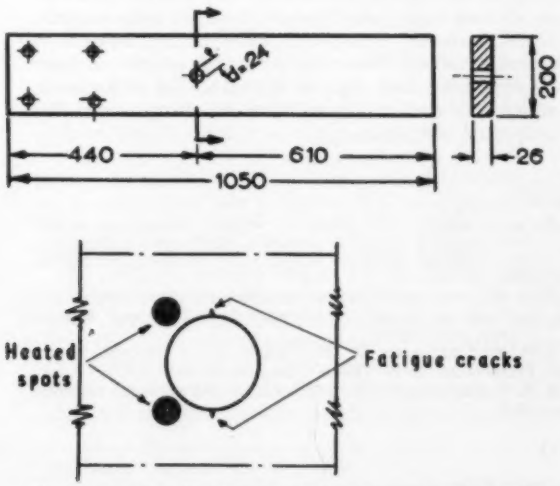


9—Test piece containing transverse butt weld and longitudinal beads

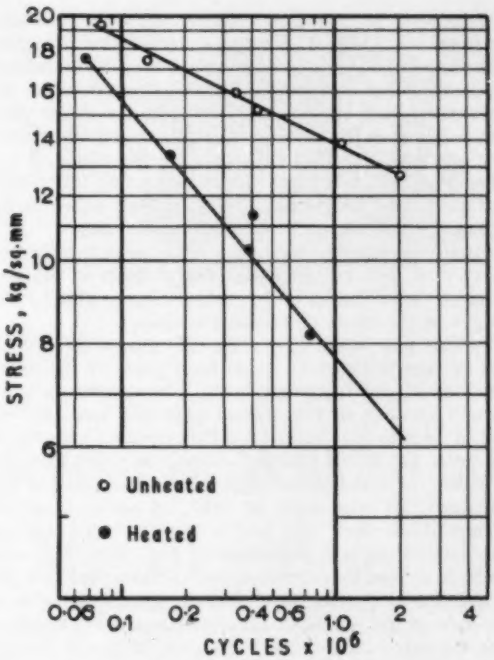
that the residual stresses in these zones were purely tensile.

That the transverse tensile residual stresses reach their maximum value along the centre of a butt weld has already been demonstrated theoretically[8] and by experiment.[3] But some explanation is required of the circumstance that maximum tensile stresses also existed at the ends of the butt weld and not compressive stresses, as should follow from the theoretical aspect of the transverse stress diagram. This diagram relates to instances where plates are welded without extension pieces, whereas the test specimens were welded with them, as in normal practice. The welding of the extension pieces (by depositing a bead on the edge) also created residual tensile stresses at the ends of the weld. X-ray measurements on a 300×26 mm test piece showed that the transverse residual tensile stresses next to the weld reached 17·2 tons/sq.in. at the end of the weld and 14 tons/sq.in. in the centre. It is thus evident that the fatigue cracks commenced in the zones containing the highest residual tensile stresses.

In test pieces of 300×26 mm cross-section, the transverse residual stress is close to the yield point of the parent metal, whilst for 70×16 mm test pieces the stress was only 1·6 tons/sq.in., as has been shown previously. The X-ray method could not be used for determining the residual stresses in test pieces with cross-sections between these two values; it was therefore impossible to trace the manner in which these stresses grew with increase in cross-section. This gap was partially filled for 200×16 and 200×26 mm



10—Special holed test pieces: (*a*) General layout; (*b*) heated spots



11—Results of tests on holed specimens

specimens by measuring the stresses by the relaxation method, which provides a sufficiently correct picture of the distribution of the stresses along the weld, although it seriously distorts the diagram of transverse stresses. Since in the present tests the residual tensile stresses at the edges of the test piece were basically created by the weld fixing the extension piece (of width 100–110 mm), and since in this distance the residual stresses were of roughly the same magnitude, it is considered that the measured stresses were close to the actual values. The residual tensile stress at the ends of 200×16 mm test pieces was 6·35 tons/sq.in., and at the ends of 200×26 mm test pieces it was 13·3 tons/sq.in.

On comparing these figures it can be seen that the transverse stresses increase with the test piece cross-section, reaching their maximum when the width is between 200 and 300 mm; this also agrees with De Garmo's experiments.

It was shown previously that the fatigue strength of a welded joint drops with increase in the tensile stresses. In consequence, the reduction in fatigue strength observed with increase in cross-section must, to a considerable degree, be related to the effect of the residual stresses, which increase with widening and thickening of the test pieces. This was confirmed by the results of tests carried out on 200×26 mm test pieces containing transverse butt welds and longitudinally deposited beads (Fig. 9).

Using small test pieces (series *C*, *D* and *E*), it was made clear that the deposition of a longitudinal bead does not alter the stress concentrations in a test piece—though the residual tensile stresses lengthwise along the test piece increase. Measurements made by X-ray have shown that the residual stresses reach

22·8 tons/sq. in. at a point against the butt weld and the longitudinal bead. Since this stress value is higher than that in similar test pieces without a longitudinal bead, a further fall in the fatigue strength must be anticipated, and the test results of Fig. 6 confirm this. The S–N curve for test pieces containing a transverse butt weld and longitudinal beads is lower than all the other S–N curves for test pieces containing transverse butt welds. The fatigue cracks commenced, as was to be expected, in the zones of highest tensile stress.

Finally, additional tests were made, which definitely prove that the basic cause of the drop in fatigue strength with increase in cross-section should be sought in the effect of residual stresses.

Special test pieces (Fig. 10), 200 mm wide and 16 and 26 mm thick, were made from plate of the same metal as in the foregoing tests. The specimens were ground smooth in the central part and were drilled with a 24 mm dia. hole. To create residual stresses at the hole, the metal was spot-heated to a temperature of 530°C. in a resistance welding machine, close to the opening; this was done on five test pieces made of 26 mm thick steel. The points at which this heating was carried out are illustrated in Fig. 10*b*. Care was taken to ensure that the structure of the metal had not changed after heating, and that the micro-hardness at the hole at the points of greatest stress concentration was the same as that at a distance from the heated points. The remaining test pieces were not heated, and there were no residual stresses in them.

All the test pieces were tested under alternating loading on a magnetic vibration machine, in the same way as those containing butt welds, but not to full rupture; they were stopped on the appearance of a small fatigue crack. The cracks in all these test pieces started at the hole at the points of highest stress concentration. The test results are graphically represented in Fig. 11, from which the following conclusions are drawn:

(i) All the points relating to test pieces with no residual stresses are sufficiently well distributed along one line, whether they are for test pieces 16 or 26 mm thick. This shows that alteration in thickness did not affect the fatigue strength of the metal. The endurance values of 200 × 16 and 200 × 26 mm test pieces were the same (8·0 tons/sq.in.).

(ii) The S–N curve for test pieces containing residual stresses is considerably lower. The fatigue strength for these test pieces, 4 tons/sq.in., is only half that for test pieces without residual stress.

(iii) Where the applied stresses are increased, the S–N curves practically coincide. This indicates once again that the influence of residual stresses increases with reduction in the magnitude of the alternating stresses.

Thus the residual stresses were, in these tests also, the basic cause of reduction in fatigue strength with increase in test piece cross-section.

CONCLUSIONS

(1) Fatigue strength values which do not reflect the real fatigue strength of actual joints under repeated loading are at present being used as the basis for design calculations. These strengths are exaggerated, since they have been obtained from tests on small test pieces which basically only show the effect of joint shape.

(2) Residual welding stresses exert a great influence on the fatigue strength of a welded joint. The effect depends on their magnitude and sign, the shape of the joint, and the magnitude of the applied stresses. As the magnitude of the applied stresses decreases, the residual tensile stresses exert greater influence and considerably lower the fatigue strength of a joint. If the magnitude of the residual stresses reaches the yield point of the parent metal the fatigue strength drops by 35–40% where the severity of notching is equal to that created by the reinforcement of a butt weld.

(3) The strength of a welded joint under repeated loading changes within wide limits depending on the dimensions of the test piece. The wider or thicker the test piece the lower is its fatigue strength. The drop in fatigue strength almost ceases above certain well-defined dimensions. The fatigue strength of butt joints becomes stable after a 26 mm thick test piece reaches a width of 200 mm.

(4) The drop in fatigue strength observed with increase in test piece dimensions must be attributed basically to the effect of residual stresses, which grow to definite values with widening and thickening of the test pieces. Other factors may evidently have an effect, but to a considerably lower degree.

(5) To obtain data which can be used for the designing of welded metal structures, it is necessary to use test pieces in which the influence of the residual stresses can develop to a high level. In this way it is possible to approach as closely as possible the actual conditions under which welded joints in real structures work. In many cases there will then be less need to resort to full size tests of joists and structures, in order to determine their fatigue strength. The sequence in which welds are made should not be disregarded when preparing test pieces.

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Welding

RESEARCH

Low Temperature Properties of Welded and Unwelded Al–5% Mg Alloy Plates

By J. E. Tomlinson, B.Sc., A.I.M., and D. R. Jackson, A.I.M.

SYNOPSIS

MECHANICAL tests have been made on the aluminium–5% magnesium alloy NP.5/6 at temperatures down to −196°C., to prove its suitability for the construction of storage tanks for liquefied gases. Its strength and resistance to crack initiation increase significantly below −75°C., although at these temperatures its resistance to crack propagation decreases somewhat and, if the specimen is large enough, a self-propagating crack will form. However, bulk yielding of material in way of the potential crack path is a prerequisite, and the work demanded is considered to give an ample margin of safety. Welds behave in a similar manner to the parent plate. The change in the form of fracture which occurs at low temperatures is briefly discussed.

SINCE 1956 serious consideration has been given to the possibility of supplementing towns' gas supplies with methane which is liberated from oil wells in various parts of the world. The only feasible method of transporting the gas is to liquefy it by refrigeration and then to bring it by sea in insulated but unpressurized and unrefrigerated tanks.

The boiling point of liquid methane is −160°C., at which temperature ferritic steels are highly brittle, and only metals that have a face-centred cubic lattice, such as aluminium, austenitic steels, and copper, can be considered for storage tanks for such a fluid.

To facilitate construction and to ensure the absence of leaks up to the highest possible stresses, the tanks would naturally be produced by welding. To reduce both the initial cost of material and the difficulties and expense of jointing, the highest-strength weldable alloy would be employed, in order to keep the metal thickness to a minimum; as far as aluminium is concerned the choice falls on the aluminium–5% magnesium alloy NP.5/6.* Whilst tensile and impact tests have been carried out at low temperatures on a wide range of aluminium alloys, notably by American workers,[1] there was no information on the behaviour of welds in the aluminium–5% magnesium alloy. The series of tests described in this paper were made to provide design data and to provide full information concerning the fracture behaviour of this material when welded.

SCOPE OF THE TESTS

In the absence of adequate supplies of liquid methane, it is not easy to cool large specimens and hold them at −160°C. A few tests were made in the range −150° to −160°C. by surrounding the specimen in the refrigerant 'Arcton 6' and cooling this with a stream of liquid nitrogen. Otherwise the tests were made in liquid nitrogen (−196°C.), in solid CO_2 (−78°C.), and at room temperature (+20°C.).

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The authors are in the Research Laboratories of the British Aluminium Company Ltd.

* BS 1477:1955, "Wrought Aluminium and Aluminium Alloys for General Engineering Purposes – Plate".

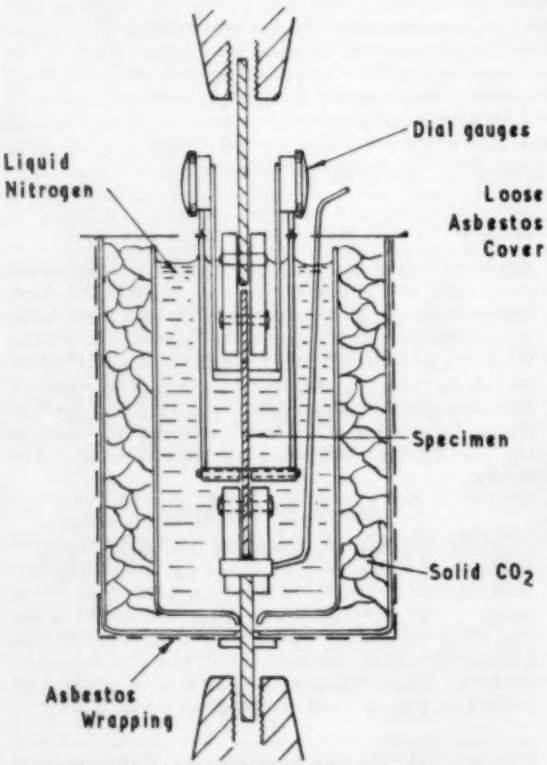
Ordinary tensile tests were carried out to provide design data, and U.S. Navy tear tests and tensile tests on centrally notched wide specimens were included to provide information on crack propagation. A number of Izod tests were made to see if this test was sensitive to any change that might occur at low temperature.

TENSILE TESTS

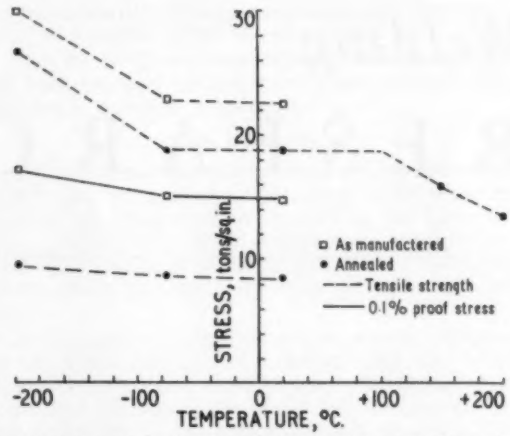
Apparatus

For the determination of tensile properties, the apparatus shown in Fig. 1 was employed. The inner vessel, containing the refrigerant, is of copper and is brazed to the lower steel shackle; the outer vessel is of mild steel held on the lower shackle by a locking ring, and the annulus between the two vessels is packed with solid CO_2, which was found to reduce heat transfer more effectively than glass wool insulation.

The extensometer was of the kind used by Phillips, Swain, and Eborall[2] at temperatures down to −78°C. For operation down to −196°C., the spindles of the dial gauges and the upper external parts of the instrument had to be coated with 'Glycol' to prevent freezing-up. The instrument was bolted to the shanks of the test piece. For the determination of proof stresses it was necessary to determine the effective gauge length; this was done by calibration against resistance wire strain gauges fixed to the gauge length of the test piece, and the effective length was found to be 3·64 in. Load/strain curves determined by strain gauges and by the extensometer were plotted, and proof stresses at +20°C. determined by the two methods were found to differ by less than 5%.



1—Apparatus for tensile tests at low temperature

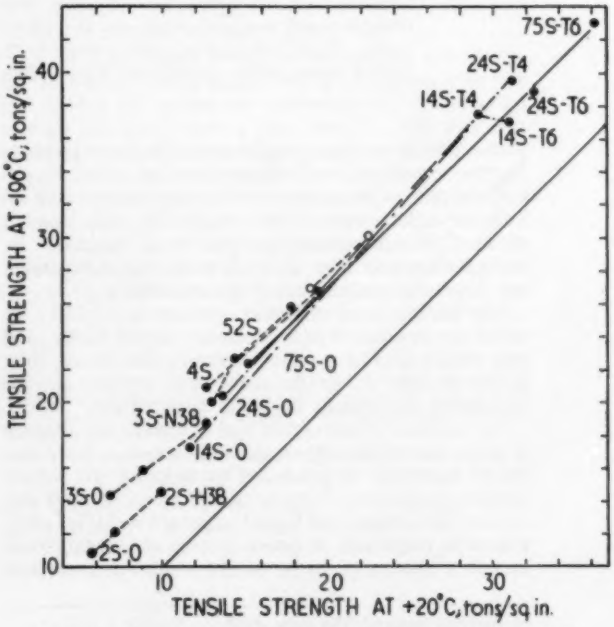


2—Variation of tensile strength and 0·1% proof stress with temperature

Material and specimens

A portion of a 3 ft × 3 ft × 0·250 in. NP.5/6 (Al–5% Mg) alloy plate taken from routine production was used to provide a series of specimens. The analysis of this plate was as follows: Si, 0·21%; Fe, 0·24%; Cu, 0·01%; Mn, 0·14%; Mg, 4·90%; Na, 0·0005%. The form of the test pieces was similar to that laid down in BS. 18, the only significant difference being a reduction of the parallel length from 2½ in. to 1½ in. However, as all elongation measurements have been referred to a gauge length of either $4\sqrt{(\text{Area})}$ or ¼ in., this difference has no bearing on the results.

Most of the specimens were tested in the 'as-manufactured' condition but some (Nos. 9, 10 and 11) were annealed at 400°C. for ½ hr. Metal in both conditions



3—Comparison of tensile strength of aluminium alloys at +20°C. and −196°C.[3]

Table I

Results of tensile tests Plain specimens to BS. 18, ½ in. wide

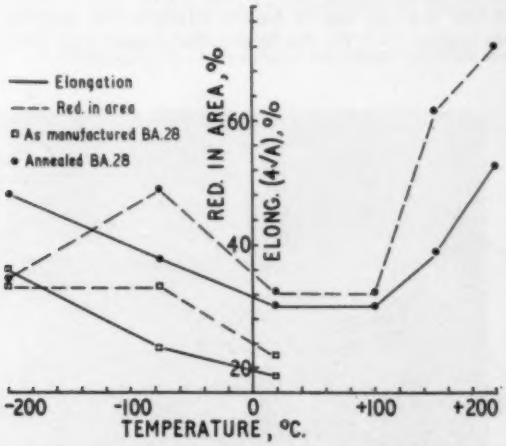
Specimen	*Direction*	*Temp., °C.*	*0·1% Proof stress*	*U.T.S.*	*Elongation, % Local*	*General*	*$4\sqrt{A}$*	*Red. in Area, %*
			tons/sq.in.					
As manufactured	*L*	+20	15·8	22·3	26	14	16·5	17
	T		14·8	22·6	32	12	19	22
	L	−78	16·3	22·8	52	12	20·5	33·5
	T		15·0	21·9	52	20	23·5	33·5
	L	−196	18·6	30·3	52	24	35	32
	T		17·2	30·0	58	32	36	33·5
Annealed at 450°C. for ½ hr	*L*	+20	8·4	19·0	56	24	30·5	31·5
	T		8·5	18·8	56	20	30·5	32·5
	L	−78	8·3	19·0	76	24	38·5	47·5
	T		8·6	18·8	68	24	38	49
	L	−196	9·2	26·9	64	40	51·5	37·5
	T		9·4	27·6	64	40	48·5	35

was tested at +20°, −78°, and −196°C., and longitudinal and transverse properties were determined in all cases. To determine the effect of stress concentrations, some specimens tested at +20°C. and −196°C. contained either Izod **V** notches, 0·020 in. deep, milled across each face, or a ⅛ in. dia. hole drilled on the centre-line of the specimen.

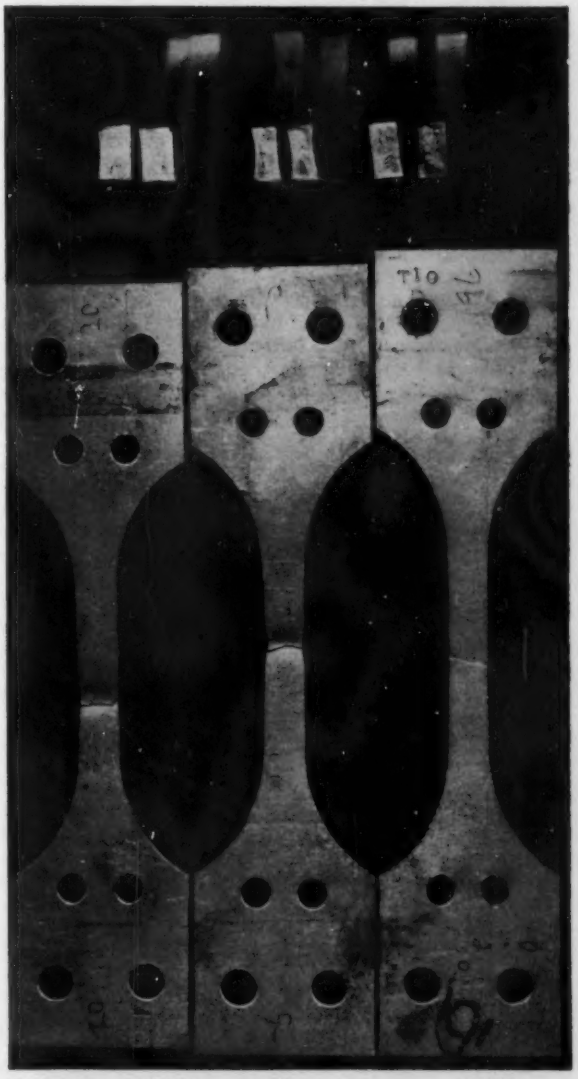
Pairs of plates, each 6 in. × 12 in. × ½ in., were butt welded with self-adjusting-arc equipment to provide specimens with longitudinal and transverse welds. The edge preparation employed was 60° included angle **V**, with a root face of one-third of the metal thickness and no root gap. Welding was carried out with automatic traverse at 24 in./min, with a current of 220 amp. After one pass was made on the grooved side, the joint was back-chipped and a sealing run was made on the reverse side using the same conditions. Specimens were tested both 'as welded' and with the weld reinforcement removed.

Plain specimens

The results of the tensile tests are given in Table I, and the tensile strength and 0·1% proof stress values are shown graphically in Fig. 2. The 0·1% proof stress increases slightly as the temperature drops to −78°C., and below this the increase becomes a little more pronounced. The tensile strength behaves similarly, except



4—Effect of temperature on elongation and reduction in area



Testing temp.: +20°C., −78°C., −196°C.

5—Fractured tensile test pieces

Table II
Tensile tests – specimens with stress raisers

Specimen	*Direction*	*Temp., °C.*	*0·1% Proof stress*	*U.T.S.*	*Elongation, % on ¼ in. Away from Fracture*	*At Fracture*
			tons/sq.in.			
Izod V notched each side to 0·020 in.	*L*	+20	—	20·3	3	16
	T		—	19·7	3	12
	L	−196	—	24·9	4	12
	T		—	25·6	4	12
¼ in. dia. hole on centre-line	*L*	+20	—	18·0	—	24
	T		—	17·9	2	28
	L	−196	—	21·4	6	32
	T		—	22·2	4	28

that below −78°C. the increase is considerably greater than with the 0·1% proof stress: in the case of annealed plate the tensile strength at −196°C. is 46·5% higher than at +20°C. However Fig. 2 shows that the increase in strength is not proportional to the strength at room temperature; both 'as-manufactured' and annealed plate increase in strength by roughly the same amount. That this is a general tendency with aluminium alloys may be shown by plotting the results of normal and low-temperature tensile tests made by the Aluminum Company of America,[3] as shown in Fig. 3. Whilst minor variations in behaviour occur between different alloys, the trend for both heat-treatable and non-heat-treatable alloys is towards a constant difference of 7 tons/sq.in. between the tensile strengths at +20°C. and at −196°C.

Ductility, as indicated by the percentage elongation on a gauge length of $4\sqrt{(\text{Area})}$, increases with decrease in temperature. However, the reduction in area at the fracture is found to increase down to a temperature in the region of −78°C., and then to decrease with further decrease in temperature, although at −196°C. it is still somewhat higher than at +20°C. The values are plotted in Fig. 4, together with the results of additional tensile tests at +150°C. and +200°C. on samples of the annealed plate. Although no tests were carried out at +100°C. in the course of this work, there is ample experimental evidence indicating the approximate constancy of both measures of ductility between +20°C. and +100°C. A view of the specimens (Fig. 5) clearly shows that, at sub-normal temperatures the general extension (that occurring before necking) increases continuously, while the elongation and reduction in area associated with the neck first increase and then gradually decrease. At slightly elevated temperatures, the elongation prior to necking decreases and the necking elongation increases significantly. This behaviour may be explained on the assumption that the 'quasi-viscous' flow of grain over grain, which occurs readily at elevated temperatures, becomes increasingly restricted as the temperature falls. On the other hand, the 'slip within grains', which occurs at lower stresses, increases as the temperature decreases, producing the combined effect illustrated in Fig. 6.

It was observed that the fractures of the specimens tested at −196°C. were principally of the mat fibrous type normal to the direction of principal stress, in contrast to the inclined 'bulk-shear' fracture characteristic of specimens tested at +20°C. (Fig. 5). Fracture was accompanied by considerable noise, and from this and the appearance of the fracture it was concluded that at −196°C., whilst the resistance of the material to crack initiation was appreciably enhanced, the resistance to crack propagation might be reduced.

Specimens with stress raisers

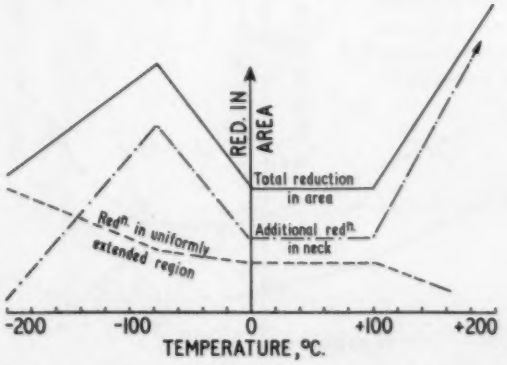
To check behaviour in the presence of stress concentrations, specimens with central holes and with Izod notches milled across their faces were tested both at +20° and at −196°C. (*see* Table II). With both forms of notch the material shows greater elongation at −196°C., both in the neighbourhood of the notch and away from it, and the nominal tensile strengths are greater at this temperature.

Specimens with transverse butt welds

The results of tests on specimens including transverse butt welds are shown in Table III. The behaviour of weld metal is much the same as that of the parent plate. Both at +20°C. and −196°C. specimens with the weld reinforcement removed have higher elongations and may have higher strength. The more sharply the reinforcement meets the plate surface, the more likely is there to be an improvement in strength when the reinforcement is removed.

Specimens of copper and F.D.P. stainless steel

For comparative purposes, specimens of two other face-centered cubic metals, copper and stainless steel, were tested at +20°C. and −196°C., and the results are given in Table IV. For both these materials, the rise in proof stress was of the same order as for NP.5/6, but the rise in tensile strength was appreciably higher: 57% for the hard-rolled copper and 116%



6—Changes in ductility with change in temperature (schematic)

Table III

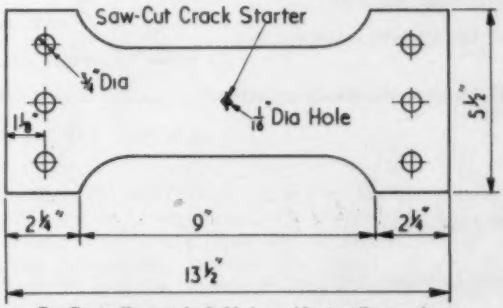
Tensile tests – specimens with butt welds

Condition	*Temp., °C.*	*Reinforcement*	*1·0% Proof stress tons/sq.in.*	*U.T.S.*	*Elong., % on $4\sqrt{A}$*
Plate 'as-manufactured', weld parallel to transverse direction	+20	on	9·5	18·4	12
		off	8·7	18·0	16·5
	−78	on	8·6	17·8	9·5
		off	—	18·5	—
	−196	on	10·25	20·45	8·5
Plate annealed, weld parallel to transverse direction	+20	on	8·3	17·9	12·5
		off	8·1	17·8	19
	−78	on	7·6	18·7	16·5
		off	7·5	17·8	19·5
	−196	on	—	24·0	14·5
		off	8·8	24·15	19·5

Table IV

Tensile tests on copper and steel

Material	*Temp., °C.*	*0·1% Proof stress tons/sq.in.*	*U.T.S.*	*Elong., % on $4\sqrt{A}$*
Copper, annealed	+20	9·73	14·36	62
	−196	—	23·1	72
Copper, cold-rolled	+20	14·15	16·5	33
	−196	16·79	25·9	71
F.D.P. stainless steel, as-rolled	+20	19·62	53·75	26
	−196	—	116·3	42



7—Centrally notched 3½ in. wide tensile specimen

for the stainless steel, compared with 46·5% for NP.5/6 The copper and stainless-steel specimens failed by inclined shear, both at +20°C. and at −196°C., and the stainless-steel specimens showed evidence of extensive 'pseudo-lamination' (*see* p. 508).

CRACK PROPAGATION TESTS

Wide centrally-notched tensile specimens

It was felt necessary to check the implication of the results of the tensile tests, that at −196°C. the resistance of NP.5/6 to crack propagation was lower than at +20°C. Accordingly, a number of specimens of the form shown in Fig. 7 were prepared for testing in tension. This test has been employed to assess the liability of high-strength heat-treatable alloys to fast fracture, and at room temperature a correlation has been established between this and the appearance of the fracture assessed on the basis of the percentage of mat fibrous as opposed to bulk shear failure. Loads to cause crack propagation and extension, around and remote from the fracture, are also measured. Plain specimens were tested together with specimens including transverse and longitudinal butt welds, prepared in the same way as for the conventional tensile tests, with samples bearing arc-strikes and samples including longitudinal butt welds containing a transverse crack. The latter were prepared by making a close-square butt joint in one pass and stopping the pass

8—Wide centrally notched tensile specimens: (*a*) With saw-cut starter cracks; (*b*) With transverse crack in longitudinal weld

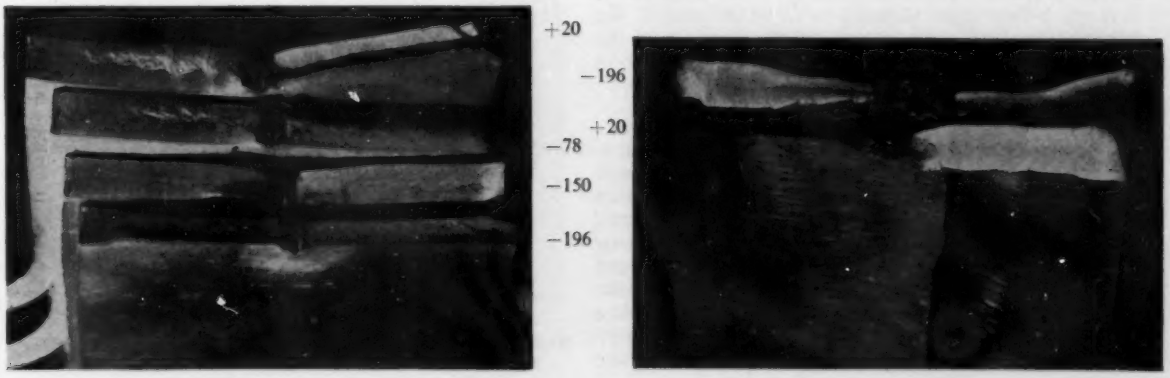


Table V
Tensile tests on wide notched specimens

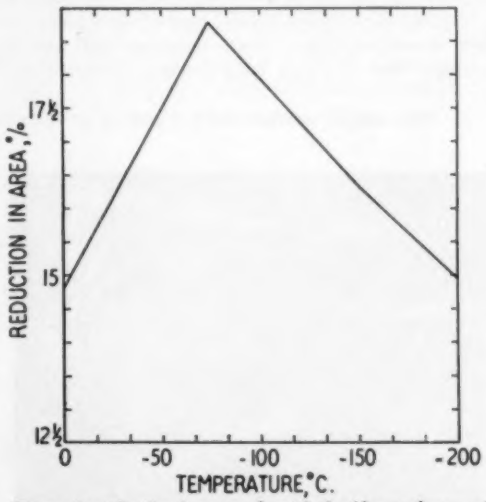
Material	*Thickness, in.*	*Direction*	*Saw-cut Length, in.*	*Temp., °C.*	*Nominal Breaking Stress, tons/sq.in.*	*General Elongation away from Fracture, %*
NP.5/6 plate	¼	*L*	¼	+20	15·3	5
		T			18·2	4
		L	¼	−78	15·5	5
		T			14·6	6
		T	¼	−150	16·6	5·5
		L	¼	−196	16·9	4·5
		T			19·8	2
NP.5/6 plate, with butt weld	¼	*L*	¼	+20	13·3	3
		T			13·8	2·5
		L	¼	−196	14·0	2·5
		T			16·2	4
NP.5/6 plate, with transverse crack in longitudinal butt weld	¼	—	—	+20	15·9	2·5
				−196	16·33	3
NP.5/6 plate with arc-melted spot	¼	—	—	+20	13·7	5
				−196	17·4	3·5
NP.5/6 plate with saw-cuts of different length	¼	*L*	¼	+20	15·1	4
				−196	15·96	5
		L	⅜	+20	11·75	2·5
				−196	13·75	2
		L	1	+20	9·87	1·5
				−196	10·37	1·5
NP.5/6 sheet	⅛	*L*	¼	+20	17·18	2
				−196	19·47	1
Pure aluminium plate	¼	*T*	1	+20	5·4	0
				−196	8·78	0
F.D.P. stainless steel	⅛	*L*	¼	+20	28·2	2
				−196	47·3	0·5
	3/16	*L*	¼	+20	33·8	—
				−196	49·8	7

mid-way along the specimen. Crater-cracks form in the crater at the stop and are not eradicated if the run is continued from directly over the crater; in fact the crack propagates up through the solidifying metal.

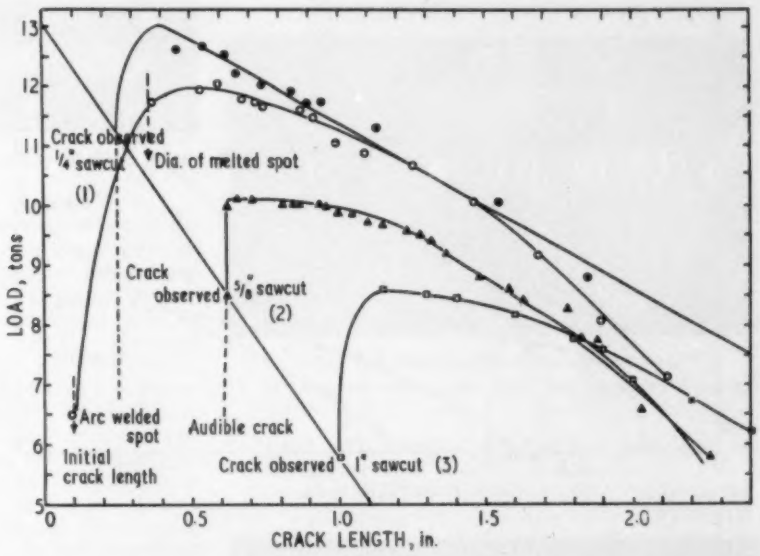
The results of these tests are given in Table V, and some typical fractures are shown in Fig. 8. It was confirmed that the fracture changes from 100% inclined bulk shear at +20°C. to virtually 100% normal mat fibrous at −196°C. At −78°C., only a minute central strip of material shows a mat fibrous fracture. At −150°C., a fine band of normal mat fibrous fracture is evident, and it is clear that the proportion of this type of fracture increases steadily as the temperature drops below approximately −78°C. A plot of the percentage reduction in area at the fracture of this type of specimen against temperature, Fig. 9, is indicative of this trend.

Propagation at +20°C.

A series of 3½ in. wide specimens of 'as-manufactured' NP.5/6 plate were prepared with central saw cuts ¼ in., ⅜ in., and 1 in. long, together with a specimen with an argon-arc melted spot containing radial shrinkage cracks at the mid-point, and these were subjected to tension. Measurements of load and crack length at the plate surface were made throughout each test, and the results are presented in Fig. 10. At a load as low as 64% of the maximum attained in a test, the formation of a minute crack at mid-thickness at the end of the saw-cut could be detected with the aid of a stethoscope. Measurement of the overall extension of the specimens (*see* Fig. 11), indicated that this minute crack formed some time after the elastic limit of the material had been exceeded. Under increasing load the

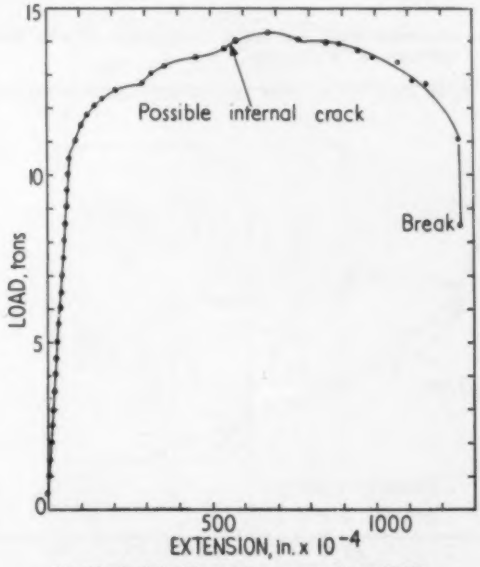


9—Change in reduction in area of notched wide specimens with change in temperature



10—Growth of cracks in NP.5/6 alloy at +20°C.

crack would first break the major surface of the specimen and then turn as it propagated, still under an increasing load. The maximum load corresponded with the condition in which the crack had just turned over into the direction of bulk shear through the full thickness of the plate. From then on, the crack grew in a series of gradually increasing steps, each corresponding with yielding of the metal in the ligaments. During this stage the load on the specimens was decreasing but the average stress on the ligaments was increasing (*see* Fig. 12). Clearly, each yield and each increment in the length of the crack occurred at high velocity. The stages of growth of the crack are indicated by the surface markings illustrated in Fig. 13, though multiple sectioning of specimens showed the actual crack front to be of the form illustrated in Fig. 14*a*. For specimens showing a 'cottage roof' or double inclined shear fracture, the crack front was as shown in Fig. 14*b*. The relationship between the length of the initial saw-cut and the maximum load withstood is shown in Fig. 15.



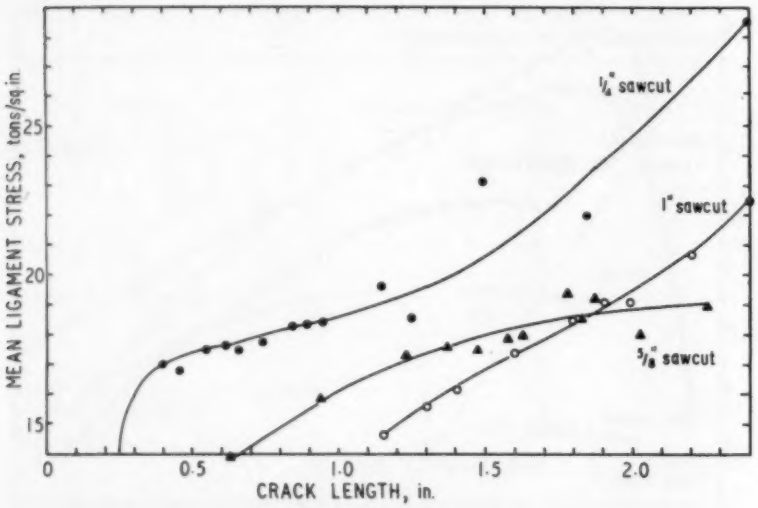
11—Typical load/extension curve at −196°C.

Propagation at −196°C.

At −196°C., the specimens failed with a normal mat fibrous fracture which was self-propagating by means of the elastic strain energy stored in the specimen. By slowing the strain rate down to 0·01 in./in./min it was possible to initiate short mat fibrous cracks, as at +20°C., and stop them from propagating, thus showing that there was a finite length for self-propagation. In this respect, NP.5/6 at −196°C. differs significantly from the recognisably brittle high-strength aluminium–zinc–magnesium alloy DTD.687A at +20°C. With fully heat-treated material of that composition, susceptible to disintegration through fast fracture, it has not proved possible to arrest cracks growing from artificial stress raisers.[4]

The fractures produced at −196°C. showed characteristic chevron markings (*see* Fig. 16), and multiple sectioning showed the crack front to be as in Fig. 17. It will be observed that the angle at which the fracture front meets the plate surface is 18°, which is exactly the value predicted from Boyd's analysis of the chevron pattern in partly brittle fractures in mild-steel plates.[5]

Specimens containing radial cracks in argon-arc melted spots behaved like the specimens with saw-cuts though the maximum loads withstood appeared to be governed by the diameter of the melted zone rather than by the visible crack length (Fig. 10). Specimens with transverse cracks in longitudinal butt welds and with central saw-cuts in transverse butt welds also had maximum loads similarly related to the crack or saw-cut length. However, both at +20°C. and −196°C., the cracks from saw-cuts in transverse welds ran off the centre-line of the weld and propagated along the edge of the weld, where there was a mechanical notch



12—
Relationship between ligament stress and crack length at +20°C.

13—Surface markings on ¾ in. thick NP.5/6 plate tested at +20°C.

arising from the sharp change in section and a metallurgical notch arising from the change in structure from cast to wrought material.

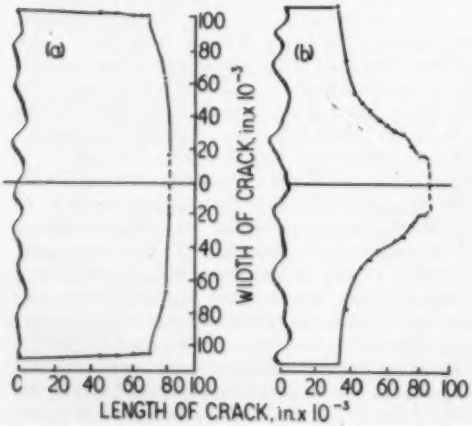
Discussion

The results of these tests on wide notched specimens confirmed that at −196°C. NP.5/6 retains a high resistance to crack initiation; furthermore, bulk yielding of all the metal in way of the potential crack path is necessary prior to propagation. The extent of this yielding will be of the order of 2·5–4·5% elongation for NP.5/6 'as manufactured', for this has been found in regions of wide test pieces remote from the crack, and the stresses required will be 5–10% higher than at +20°C. However, when the crack has been started it will grow slowly to only a very small extent before propagating as a fast fracture. It is important to appreciate that a weld crack will not simply propagate as a fast fracture; bulk yielding and the growth of the weld crack into yielded metal are first necessary.

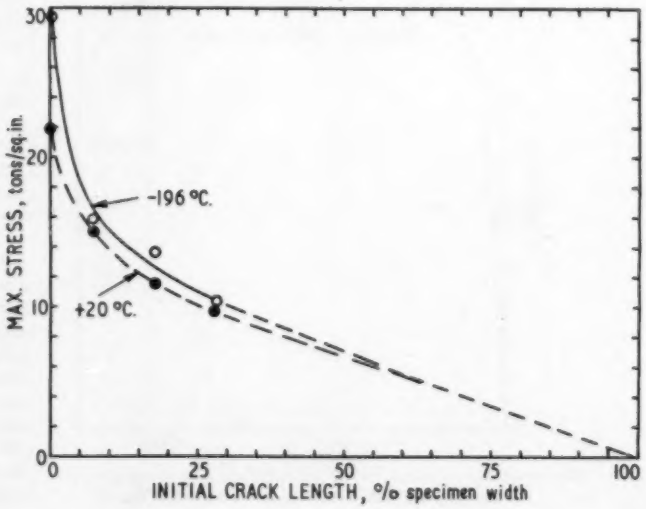
Comparative tests

Similar tests on ⅛ in. and 3/16 in. F.D.P. titanium-stabilized 18% Cr–8% Ni steel gave similar results,

14—Shape of crack front; (*a*) Inclined shear fracture in full thickness as in specimen shown in Fig. 13; (*b*) 'cottage roof' type of fracture

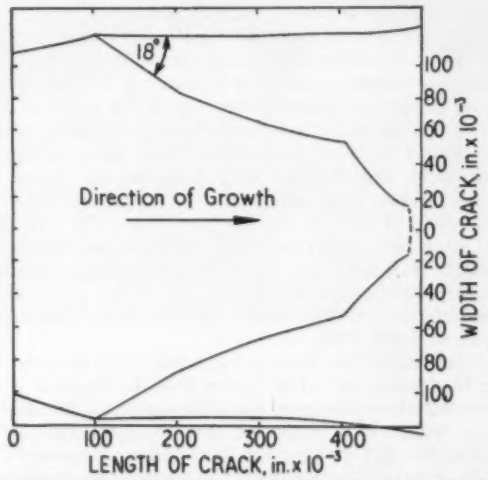


15—Maximum stress and initial crack length of wide tensile specimens of NP.5/6 alloy

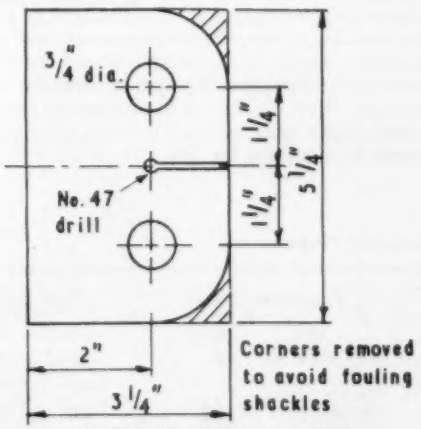


16—Typical fracture in ¼ in. thick NP.5/6 plate showing chevron markings

the general features of which are summarized at the end of Table V. The fracture changed from 100% inclined shear at +20°C. to a mainly normal mat fibrous surface at −196°C., and from slow tearing to self-propagation. The ratio of (maximum load at −196°C.)/(maximum load at +20°C.) is much greater for F.D.P. steel than for NP.5/6, *e.g.*, 1·68 for ¼ in. cracks in ⅛ in. steel plate, and 1·64 for ¼ in. cracks in $\frac{3}{16}$ in. steel plate, compared with 1·1 for ¼ in. cracks in ¼ in. NP.5/6 plate. As far as elongation remote from the starter crack was concerned, the stainless-steel plates showed marked variability, values from ½% to 7% being recorded on specimens tested at −196°C. However, the general trend of the results shows that at



17—Shape of crack front in ¼ in. thick NP.5/6 plate tested at −196°C.



18—U.S. Navy tear test specimen

Crack runs into overlap

Crack runs along middle of flush-ground weld

19—Fractures in welded U.S. Navy tear specimens tested at +20°C.

very low temperatures the fracture behaviour of F.D.P. stainless steel and NP.5/6 is similar.

Annealed aluminium of commercial purity in the form of ¼ in. plate was tested with a 1 in. long central saw-cut, and showed 100% inclined shear failure at both +20°C. and −196°C. At both temperatures the crack was slow-running.

U.S. Navy tear tests

The tests on wide, notched tensile specimens had shown that fast, self-propagating fracture could occur in NP.5/6 at temperatures below −78°C., and it was of interest to determine the energies required to initiate and propagate fractures at +20°C. and at low temperatures. In ref. 66 in the review by Bogardus, Stickley and Howell,[1] the results of tear tests on an alloy of the H.20 type showed a marked reduction in the energy to propagate fractures even at −78°C., particularly in the longitudinal direction, and it was felt necessary to check if this applied also to NP.5/6.

Specimens for the U.S. Navy tear test[6] were cut from 0·4 in., ¼ in., and $\frac{3}{16}$ in. plates to the pattern shown in Fig. 18. During the course of the test a load is applied through hardened steel pins bearing on the ¾ in. dia. holes; material at the root of the slot is in tension but material behind the slot is first in compression and then in tension as the crack grows. The results are given in Table VI.

Though the form of the fracture changed as in the other tests, none of the specimens tested at low temperatures failed by fast fracture. Since the centrally notched wide tensile specimens did break in this way, it is evident that fast fracture does not occur in the tear test pieces because it is not possible to store sufficient strain energy in small specimens of this kind. These results demonstrate that a fracture showing chevron markings, already recognized as not necessarily involving cleavage, is not necessarily characteristic of an unstable condition involving a self-propagating crack, and that this feature depends upon the configuration of the test assembly.

The energy to initiate a crack invariably rises with decrease in temperature, and it is suggested that this is primarily a reflection of the higher elastic strength of the material and the increased elongation up to the onset of necking. The energy to propagate a crack may decrease, and this is in agreement with the results of the tests on wide notched specimens. However, the

Table VI

U.S. Navy tear tests

Material	*Max. Load, tons*		*Initiation Energy, ft–lb*		*Propagation Energy, ft–lb*		*Total Energy to Tear, ft–lb*	
	+20°C.	*−196°C.*	*+20°C.*	*−196°C.*	*+20°C.*	*−196°C.*	*+20°C.*	*−196°C.*
NP.5/6, ¼ in.	3·18	4·08	71·6	137·8	120·5	98·5	192·1	236·3
As above, butt-welded	2·55	3·0						
NP.5/6, $\frac{3}{16}$ in.	2·00	2·35	70·1	97·0	62·8	93·5	162·4	160·9
NP.5/6, ½ in.	3·2	3·75	93·3	102·6	121·8	80·0	215·1	182·6
As above, butt-welded	2·8	3·45	52·6	79·6	29·8	51·6	82·4	131·2
NP.5/6, 0·4 in.	5·28	5·8	172·0	197·0	163·0	121·0	335·0	318·0
As above, butt-welded, dressed flush	4·22	4·8	107·5	125·0	73·4	61·5	232·9	186·5

decrease is not of great proportions, and is much less than that recorded by Bogardus, Stickley, and Howell.[1] It is possible that the difference can be attributed to the difference in thickness of the specimens (¼ in. *versus* ¾ in.), but this seems unlikely to be the full explanation.

The behaviour of the welded specimens of ¼ in. NP.5/6 plate shows that a stress concentration which can arise with excessive reinforcement on a badly shaped underbead, as in these specimens (Fig. 19), can have a marked effect on the energy required for crack propagation. Furthermore, there is an indication that this effect is no greater at −196°C. than at +20°C., and may be less.

The maximum loads withstood by the specimens tested at −196°C. were from 10 to 28% higher than for specimens tested at +20°C., and therefore in line with the results of the tensile tests. It is of interest that while efficiency (measured in terms of the maximum load withstood relative to that for unwelded plate) of the dressed weld in 0·4 in. NP.5/6 plate is 80% at +20°C. and 83% at −196°C., compared with 87·5% and 92% respectively for the badly shaped and undressed weld in ¼ in. plate, the propagation energy at +20°C. is appreciably higher, even taking into account, at a proportional rate, the difference between the thickness of the specimens. The comparison of the fracture energies for the ¼ in. and 0·4 in. plates made in Table VII indicates that energies are substantially proportional to the thickness.

Note on fracture appearance

The normal mat fibrous fractures of centrally notched wide specimens and tear specimens tested at −196°C. frequently had a laminated appearance, as shown in Fig. 20*a*. This applied to fractures in F.D.P. stainless steel as well as in NP.4, NS.5, and NP.5/6. Figures 20*b* and *c* show that there is no lamination in the usual sense, but that bands of the metal have fractured as individual tensile test pieces.

IZOD IMPACT TESTS

Izod impact tests have been made on a number of samples of plate over ½ in. thick. Square, three-notch specimens of standard dimensions were tested at +20°C. and −196°C. In some cases both longitudinal (*L*) and transverse (*T*) specimens were available; in others, specimens could only be taken in one direction. In all instances, notches were cut in the plane of the major face of the plate (*F*) and through the thickness of the plate (*N*). Where specimens were taken across a butt weld, the notches were disposed as shown in Fig. 21. Tests were made on a 1 in. plate of NP.4, a ⅝ in. plate of NP.5/6, and a 1 in. plate of NP.5/6 containing a butt weld made with the inert-gas metal-arc process. The joint was made in three passes, the edge preparation being a double 90° **V** with a ⅛ in. nose. The analyses of the plates and the Izod values are given in Table VIII.

It is evident that there is a general slight decrease in the Izod values at −196°C. and that the decrease is of approximately the same proportions in all the directions tested. It is suggested that these changes arise mainly through a reduction in the energy required for crack propagation. Values in the longitudinal direction are markedly superior to those in the transverse direction mainly, it is suggested, because of the tendency of the grains to be elongated in the direction of rolling, and because the presence of micro-porosity and strings of insoluble constituents (which would have little effect on longitudinal ductility) can be expected to have a deleterious effect on flow in the transverse direction. The latter factor may account for the relatively low transverse Izod values for the ⅝ in. thick NP.5/6 of

Table VII

Fracture energies for 0·25 in. and 0·4 in. NP.5/6 plates

Plate	*Initiation Energy, ft-lb*		*Propagation Energy, ft-lb*		*Total Energy, ft-lb*	
	+20°C.	*−196°C.*	*+20°C.*	*−196°C.*	*+20°C.*	*−196°C.*
NP.5/6 ¼ in. plate, actual	93·3	102·6	121·8	80	215·1	182·6
Value for ¼ in. plate calculated from results on 0·4 in. plate of NP.5/6	107·5	123	102·5	75·5	210	198·5

(a)

20—
Form of fractures produced at −196°C.

(*a*) Wide tensile specimen showing 'lamination'

(*b*) Section of normal mat fibrous fracture showing 'cup and cone' failure of individual grain with intergranular boundaries. Etched ×1000

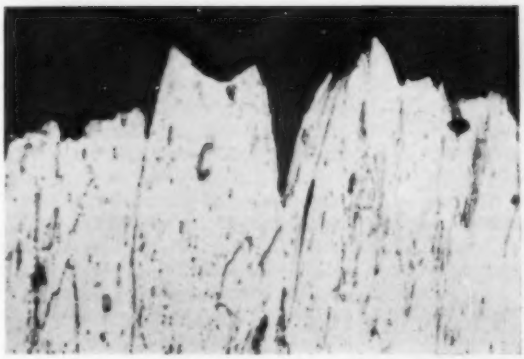
(*c*) Different part of fracture in (*b*) showing 'cup and cone' failure of group of grains with intergranular boundaries. Etched ×1000

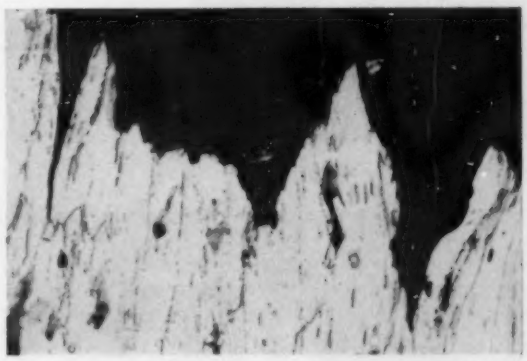
special composition in both the face and normal directions. The impact strength of weld metal is somewhat lower than that of the parent plate, though it is no more seriously affected than the latter by exposure to low temperatures.

CONCLUSIONS

It is evident that the principal effect of very low temperatures on the tensile properties of this material is to prolong the process of general elongation, *i.e.*, to delay the onset of the instability which occurs in tensile loading, and which leads to necking and fracture. Since annealed material is therefore more intensely work hardened at the onset of failure, less ductility will be in evidence in the neighbourhood of the fracture, and in particular the neck will be of such small dimensions that bulk shear across it is impossible. Hence normal, mat fibrous fractures will result, though such fractures will only be produced after extensive general plastic deformation of the structure. This is in marked contrast to the behaviour of body-centered cubic materials, such as ferritic steels, and

(b)



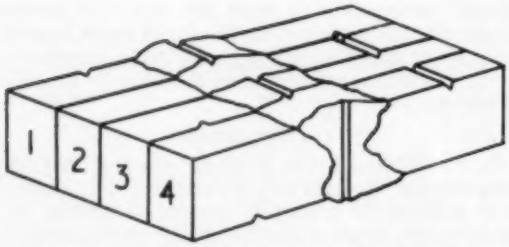


(c)

Table VIII
Izod tests

Material	*Temp., °C.*	*Plate 1* *Direction of Notch*	*Izod Values, ft–lb*	*Weld* *Direction of Notch*	*Izod Values, ft–lb*	*Plate 2* *Direction of Notch*	*Izod Values, ft–lb*
NP.5/6, 1 in. thick; Cu, 0·04%; Mg, 5·20%; Si, 0·15%; Fe, 0·29%; Mn, 0·43%	+20	*NT*	14·5, 15	*F*	12·0, 11·5	*NT*	14·5, 14
		FT	18·0, 17·5	*T*	11·0, 10·0	*FT*	17·0, 17
		NT	14·5, 14·5	*R*	10·5, 11·0	*NT*	14·5, 15
	−196	*NT*	11·9, 11·2	*F*	7·6, 8·7	*NT*	11·0, 11·6
		FT	11·2, 14·9	*T*	10·4, 8·4	*FT*	13·0, 12·5
		NT	12·5, 11·8	*R*	9·8, 7·7	*NT*	11·0, 11·8

Material	*Temp., °C.*	*Plate* *Longitudinal* *Direction of Notch*	*Izod Values, ft–lb*	*Transverse* *Direction of Notch*	*Izod Values, ft–lb*
NP.5/6, ½ in. thick; Cu, 0·01%; Mg, 5·25%; Si, 0·14%; Fe, 0·30%; Mn, 0·44%	+20	*F*	27·0, 27·0	*F*	19·5, 20·5
		N	21·0	*N*	16·0
	−196	*F*	20·8	*F*	15·4
		N	13·5, 14·5	*N*	11·4, 9·0
NP.5/6, ⅝ in. plate; Cu, 0·03%; Mg, 4·10%; Si, 0·16%; Fe, 0·30%; Mn, 0·70%	+20	*F*	34·5, 33, 34, 35, 34·5, 35	*F*	20, 15, 19·5, 20, 20·5, 19·5
		N	21, 21, 22	*N*	15, 14, 12·5
	−196	*F*	32·5, 32·5, 33, 33, 33, 32	*F*	20, 20·5, 19, 20, 18·0, 18·5
		N	21·5, 20, 20	*N*	15·5, 12·5, 13
NP.4, 1 in. plate; Cu, 0·01%; Mg, 2·15%; Si, 0·19%; Fe, 0·43%; Mn, 0·40%	+20	*F*	48, 60, 41, 47·5, 43, 43	*F*	36, 36·5, 40·5, 37, 35, 38
		N	45·5, 42·5, 39·5	*N*	30, 37, 30
	−196	*F*	38, 43, 40, 38, 39, 41·5	*F*	23, 32, 30·5, 34, 32, 27·5
		N	34·5, 36, 35	*N*	30, 25, 26



21—Arrangement of Izod specimens in butt welds
1 and 3 : face of weld
2 and 4 : through weld
1 and 2 : room temp. test
3 and 4 : low temp. test

offers a useful factor of safety. Mat fibrous fractures in the aluminium alloy may show forms of chevron markings, but these are not necessarily indicative of a fast running crack. Cracks of such an appearance produced in wide notched tensile test pieces are self-propagating, but cracks in the smaller tear test specimens are not. The results of these and other tests on a range of aluminium alloys indicate that the risk of the formation of self-propagating cracks may be determined by the severity of any stress concentrations that may exist. The more severe the stress concentration, the less is the risk of the crack being self-propagating simply because the material yields and begins to tear under lower loads and hence in a less intensely work hardened state.

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2. V. A. Phillips, A. J. Swain, and R. Eborall: *J. Inst. Metals*, 1953, vol. 20, pp. 625–647.
3. Reference 61 in (1) above. Unpublished work by Aluminium Research Laboratories, Aluminum Company of America.
4. Unpublished work by J. E. Tomlinson.
5. G. M. Boyd: *Engineering*, 1953, vol. 175, Jan. 16, p. 65.
6. N. A. Kahn and E. A. Imbembo: *Welding J.*, 1948, vol. 27, pp. 169s–186s.

Extruded Aluminium Alloys for Low Temperature Service

AN ASSESSMENT OF THE SUITABILITY OF TWO MATERIALS

By R. J. Durham

SYNOPSIS

To assist in determining the suitability of two aluminium alloys, Noral 50SWP (HE9) and Noral 54SM (NE5), for service at low temperatures, tensile tests and U.S. Navy tear tests (which indicate the resistance of the material to the propagation of cracks) were carried out at room temperature and at −195°C. Both welded and weld-free test pieces made from flat extrusions $\frac{5}{16}$ in. thick in both alloys and extruded 10 in. dia. pipe in Noral 50SWP were tested. The energies absorbed by each tear test specimen up to and subsequent to the incidence of the maximum load were measured. It was found that the tensile strength had increased at the low temperature without significant loss in elongation and that in no instance were cracks more readily formed and propagated at −195°C. than they were at room temperature.

In recent years considerable interest has been shown in materials that might be suitable for structural purposes at very low temperatures. The known brittle behaviour of many steels at temperatures below about 0°C. means that the superior performance of alternative materials is of considerable importance.

The U.S. Navy tear test is a simple and convenient test for determining whether a material behaves in a brittle or ductile manner. It was therefore used to study the properties at 20°C. and −195°C. of Noral 54SM (BS 1476:NE.5.M) and Noral 50SWP (BS 1476: HE.9.WP). The properties of both welded and unwelded materials were measured.

MATERIALS AND PREPARATION OF SPECIMENS

Specimens of 5 in. × $\frac{5}{16}$ in. extruded sections of both 50S and 54S alloys were used for the tests. Some 10 in.

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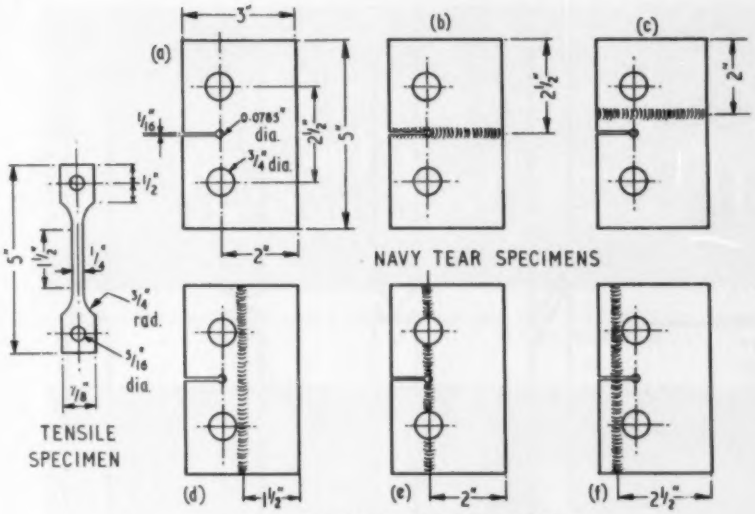
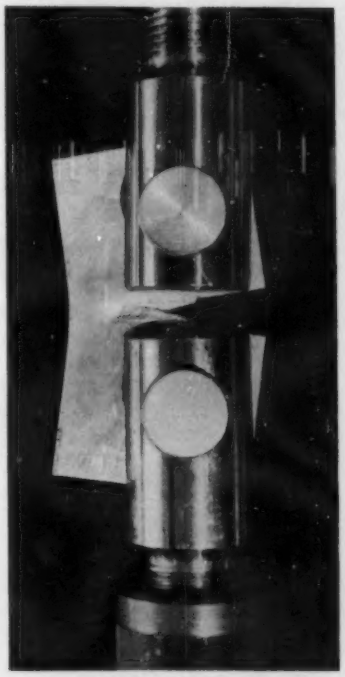
The work described was carried out at the request of the Northern Aluminium Company Ltd.



1—Test pieces and positions of welds



2—Loading shackles

dia. porthole extruded pipe of 50S of approximately $\frac{5}{16}$ in. wall thickness was also tested. The compositions of these materials are given in Table I.

Tensile and Navy tear test pieces, both longitudinal and transverse, were prepared from the parent materials and from welded joints. The welds were made with a procedure suitable for welding pipe. The edges to be joined were given a J preparation and the welds were made in two passes, using the argon-shielded tungsten-arc process and N6 filler wire; a steel backing bar with a radiused groove was used during welding.

The welded transverse test pieces were cut from the flat $\frac{5}{16}$ in. × 5 in. extrusions, which had been slit lengthways into two pieces and then welded together

Table I
Compositions of test materials

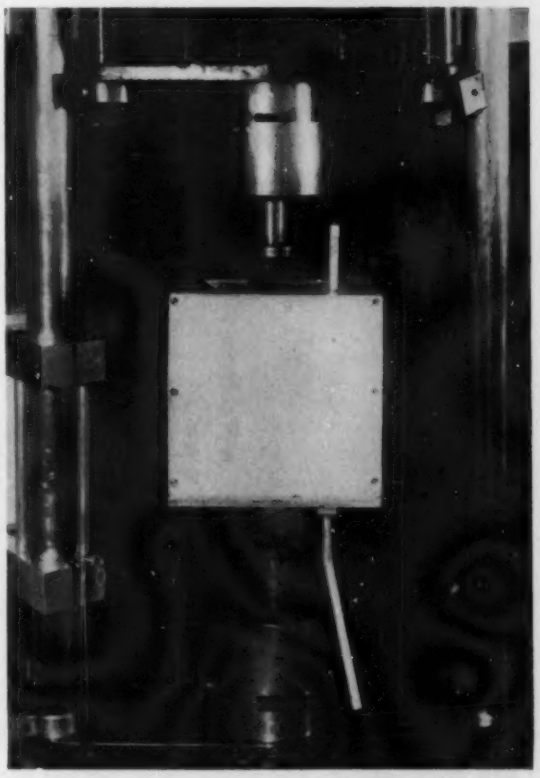
Material	*Cu,%*	*Mg,%*	*Si,%*	*Fe,%*	*Mn,%*	*Ti,%*
54S section	0·029	3·34	0·22	0·30	0·57	0·010
50S section	0·028	0·65	0·47	0·34	0·035	0·010
50S pipe	0·021	0·63	0·50	0·40	0·013	0·006

along the slit edges. The welded longitudinal test pieces were similarly obtained from short lengths of extrusion and flattened pipe that had been welded transversely. The flattened pieces of pipe were made from short lengths of unheat-treated pipe, which were slit longitudinally, flattened, solution heat-treated, stretched 1% in the direction parallel to the original pipe axis and finally precipitation heat-treated.

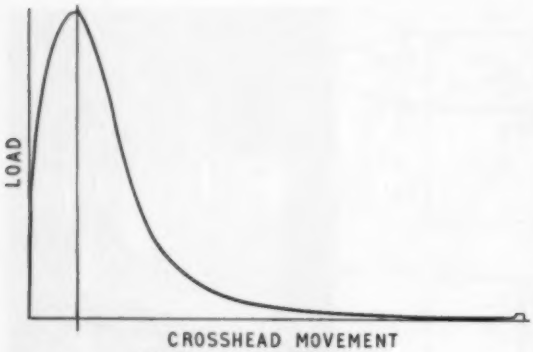
The dimensions of the tensile test pieces and the Navy tear test pieces, and the weld positions are shown in Fig. 1. In the welded tear specimens the welds were so placed that tears parallel to and normal to the welds could be produced.

TESTING PROCEDURE

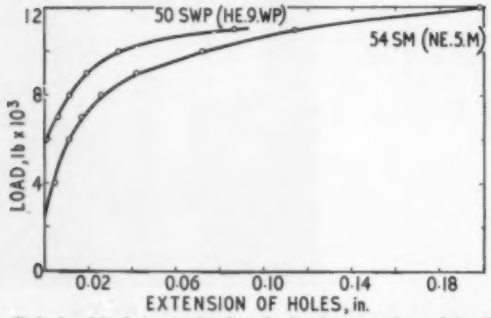
Both the tensile test pieces and the Navy tear test pieces were loaded through shackles, shown in Fig. 2,



3—Liquid air bath surrounding Navy tear test specimen



4—Typical autographic record



5—Relationship between load and plastic extension of loading holes in Navy tear test pieces

in a hydraulic testing machine of 30,000 lb maximum capacity. To carry out tests at −195°C. the specimens were surrounded by liquid air contained in a small insulated bath supported on the lower shackle (*see* Fig. 3). Provision was made for emptying the bath between tests.

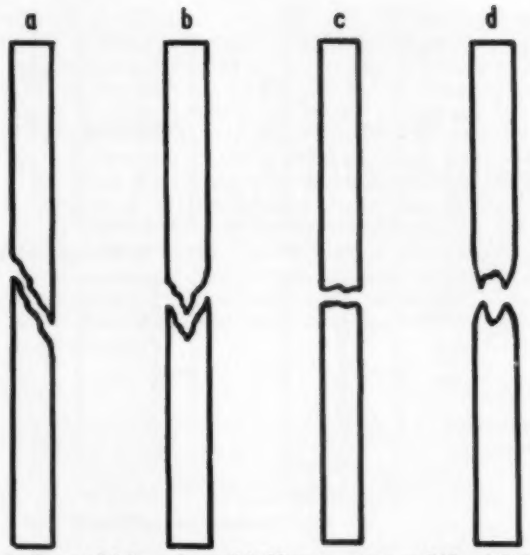
During the first few tests at −195°C. the temperatures of the specimens were measured with copper–constantan thermocouples. It was found that each specimen cooled quickly to −195°C. and remained at that temperature with but slight variation, and no temperatures were measured in subsequent tests.

Navy tear tests

The Navy tear test was devised in 1948 by Kahn and Imbembo* to determine on small test pieces the temperature of transition from ductile to brittle behaviour of certain steels, and to compare these results with those from large specimens. Simple, easily made test pieces are used, and the energies needed to start and propagate a tear are measured. These are determined from the areas under the load/extension diagram on either side of the ordinate at the maximum load.

In the series of tests described, the loads and crosshead movements were plotted autographically by a mechanically-operated drum recorder. An example of the diagrams so obtained is shown in Fig. 4. It was found necessary to apply two corrections to the energy values determined from the diagrams because the crosshead movements did not accurately represent the extensions of the specimens between the loading holes.

* N. A. Kahn and E. A. Imbembo: *Welding J.*, 1948, vol. 27, pp. 169*s*–182*s*.



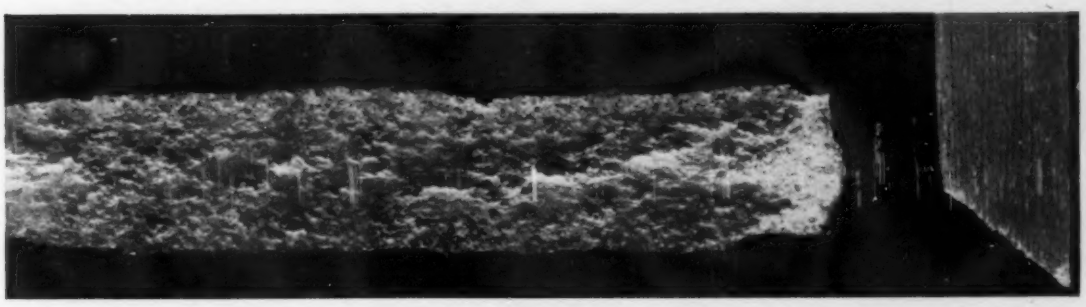
6—Types of failure observed in Navy tear tests on 54S and 50S extruded materials: (*a*) Type 1—failure in a single oblique plane; (*b*) Type 2—failure in two oblique planes; (*c*) Type 3—Failure normal to the plane of the materials; (*d*) Type 4—Failure with a double cup appearance

The first correction compensated for the deformation of the testing machine under load. Deformations at various loads were measured and hence the energy stored in the testing machine was determined. The appropriate values were then deducted from the energy necessary to start the tear (as shown by the recorder graph) and added to the propagating energy

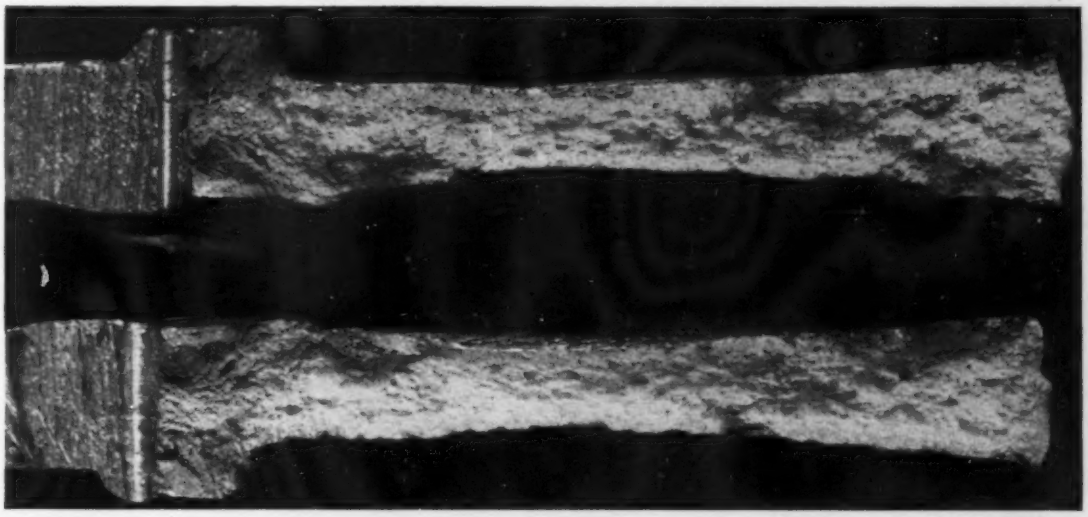
Table II

Tensile properties

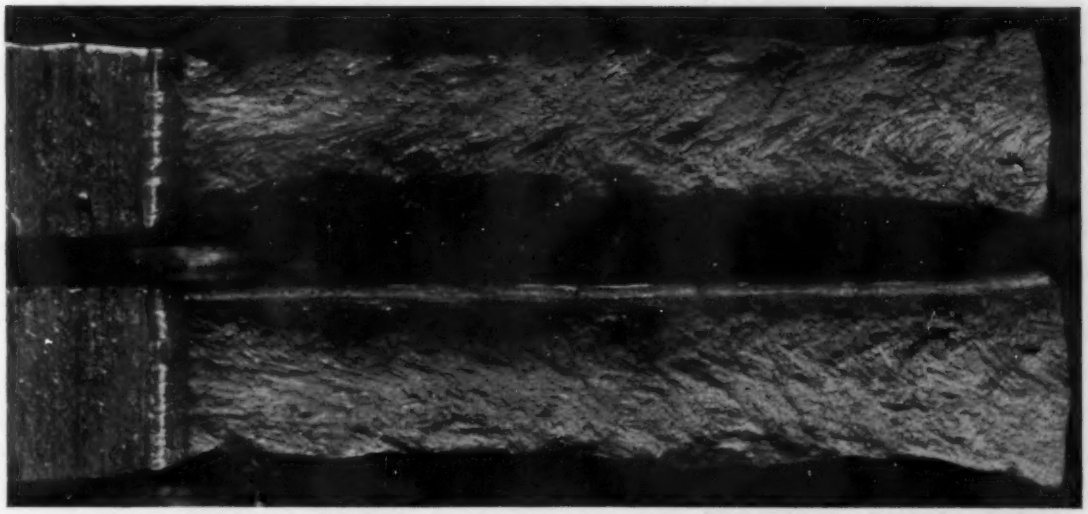
Specimen	*Grain Direction*	*Test Temp., °C.*	*Proof Stress, tons/sq.in. 0·1%*	*0·2%*	*U.T.S., tons/ sq.in*	*Elong. on 1 in., % ($4\sqrt{A}$)*
50SWP extrusion (HE.9.WP)						
Unwelded	longitudinal	Room	12·5	12·7	14·6	28
	,,	−195	—	—	19·95	35
	transverse	Room	11·1	11·95	14·4	26½
	,,	−195	—	—	19·8	29
Butt welded	longitudinal	Room	4·8	5·4	8·7	18
	,,	−195	—	—	14·7	26
	transverse	Room	4·95	5·35	8·9	21
	,,	−195	—	—	15·5	25
50SWP pipe (HV.9.WP)						
Unwelded	longitudinal	Room	12·7	—	15·4	25
	,,	−195	—	—	21·7	22
	transverse	Room	10·65	11·3	15·1	21
	,,	−195	—	—	20·8	21
Butt welded	longitudinal	Room	5·1	5·6	9·65	16½
	,,	−195	—	—	15·05	22
54SM extrusion (NE.5.M)						
Unwelded	longitudinal	Room	12·0	12·05	18·05	22½
	,,	−195	—	—	27·1	33
	transverse	Room	10·65	11·0	17·7	29
	,,	−195	—	—	25·7	34
Butt welded	longitudinal	Room	7·45	8·1	17·3	15
	,,	−195	—	—	23·7	18
	transverse	Room	7·0	7·6	17·0	25
	,,	−195	—	—	22·8	18



7—Failure along one oblique plane in transverse test piece type *a* of 50SWP pipe material. Tear occurred along porthole join; tested at room temperature ×3



8—Failure along two oblique planes in transverse test piece type *e* of 50SWP pipe material. Tear commenced in centre of weld; tested at −195°C. ×3



9—Failure in plane normal to plane of specimen in transverse test piece type *b* of 54SM extrusion. Tear occurred along the weld; tested at −195°C. ×3

value. The second correction took into account the work done in extending the loading holes in the test piece. Unslotted test pieces of each material were loaded incrementally, the load being removed and the loading hole diameters being measured between successive loadings. Graphs of load against plastic deformation were thus obtained (Fig. 5). Hence it was possible in each test, knowing the maximum load and the permanent extensions of the loading holes, to estimate the amount of work done in producing these extensions.

In their original paper Kahn and Imbembo did not make these corrections, and in comparing test results from different sources it is necessary to determine whether or not such corrections have been made.

RESULTS

Tensile tests

The tensile properties of welded and unwelded specimens are shown in Table II. All the materials showed a considerable increase in tensile strength if cooled to −195°C., without loss in elongation. The

Table III

Results of Navy tear tests on 50SWP (H.9) extruded section and pipe and on 54SM (N.5) extruded section

Type of Test Piece (see Fig. 1)	*Grain Direction*	*Test Temp., °C.*	*Maximum Load, lb*	*Work done on test piece,* ft.lb*						*Fracture appearance† (see Fig. 6)*
				To start tear		*To propagate tear*		*Total*		
				Gross	*Corrected*	*Gross*	*Corrected*	*Gross*	*Corrected*	
50SWP extrusion (HE.9.WP)										
a	Longitudinal	Room	8,600	56	34	149	153	205	187	4
a	,,	−195	11,130	93	52	166	173	259	225	3
a	Transverse	Room	9,000	69	38	122	125	191	163	1 and 2
a	,,	−195	11,180	86	40	136	144	222	184	1
b	Longitudinal	Room	6,600	53	29	114	117	167	146	1
b	,,	−195	7,540	42	26	80	82	122	108	3
b	Transverse	Room	6,550	67	29	121	123	188	152	1
b	,,	−195	7,200	53	27	71	74	124	101	3
c	Longitudinal	Room	6,050	79	41	121	123	200	164	4
c	,,	−195	8,450	109	48	196	200	305	248	4
c	Transverse	Room	6,100	54	27	128	130	182	157	4
c	,,	−195	8,750	108	44	193	197	301	241	3
d	Longitudinal	Room	7,040	170	58	161	163	331	221	4
d	,,	−195	9,950	285	79	199	203	484	282	4
d	Transverse	Room	6,390	110	41	120	123	230	164	1
d	,,	−195	8,070	149	55	159	162	308	217	3 and 4
e	Longitudinal	Room	5,850	43	6	172	174	215	180	4
e	,,	−195	7,900	94	19	211	215	305	234	4
e	Transverse	Room	5,610	48	7	173	175	221	182	1 and 4
e	,,	−195	7,170	65	8	226	229	291	237	1 and 2
f	Longitudinal	Room	7,535	143	57	159	162	302	219	4
f	,,	−195	10,270	215	94	201	206	416	300	4
f	Transverse	Room	6,840	103	24	164	167	267	191	4 and 1
f	,,	−195	10,250	215	67	209	214	424	281	1 and 2
50SWP pipe (HV.9.WP)										
a	Longitudinal	Room	8,780	58	32	125	129	183	161	1 and 2
a	,,	−195	10,630	58	28	123	128	181	156	3
a	Transverse	Room	8,660	50	31	75	78	125	109	1
a	,,	−195	10,610	54	28	90	95	144	123	1
a‡	Longitudinal	Room	8,710	65	32	117	120	183	152	1 and 2
a‡	,,	−195	10,540	75	41	129	134	204	175	1 and 2
a§	Transverse	Room	8,880	67	31	81	105	148	136	1
a§	,,	−195	10,730	54	23	88	93	142	116	2 and 3
b	Longitudinal	Room	6,000	50	27	137	138	187	165	1
b	,,	−195	7,750	48	29	96	99	144	128	3
c	Longitudinal	Room	5,820	56	32	130	132	186	164	4
c	,,	−195	8,500	92	53	213	216	305	269	4
d	Transverse	Room	7,700	167	57	98	101	265	158	1 and 2
d	,,	−195	9,950	177	56	190	194	367	250	3 and 2
e	Transverse	Room	5,670	60	15	142	143	202	158	3 and 4
e	,,	−195	8,050	75	9	183	186	258	195	2

* Usually the gross values only are quoted (*see* ref. p. 512)
‡ Tear began at and ran normal to porthole join
§ Tear began in and ran along porthole join

† Fracture appearance classifications:
1. Failure in a single oblique plane
2. Failure in two oblique planes
3. Failure normal to the plane of the material
4. Failure with a double cup appearance

Table III (*continued*)

Results of Navy tear tests on 50SWP (H.9) extruded section and pipe and on 54SM (N.5) extruded section

Type of Test Piece (*see Fig. 1*)	*Grain Direction*	*Test Temp., °C.*	*Maximum Load, lb*	*Work done on test piece,* ft.lb*						*Fracture appearance†* (*see Fig. 6*)
				To start tear		*To propagate tear*		*Total*		
				Gross	*Corrected*	*Gross*	*Corrected*	*Gross*	*Corrected*	
54SM extrusion (NE.5.M)										
a	Longitudinal	Room	9,400	91	46	148	156	239	202	1
a	,,	−195	11,870	149	62	156	138	305	200	1
a	Transverse	Room	9,330	91	44	92	96	183	140	1
a	,,	−195	11,350	127	46	117	120	244	166	3
b	Longitudinal	Room	8,150	83	23	128	131	211	154	1
b	,,	−195	9,850	111	28	95	100	206	128	3
b	Transverse	Room	8,350	132	30	118	121	250	151	1
b	,,	−195	9,975	141	33	105	119	246	152	3
c	Longitudinal	Room	7,820	79	32	121	124	200	156	1
c	,,	−195	10,200	139	46	164	169	303	215	1
c	Transverse	Room	7,810	97	37	99	102	196	139	1
c	,,	−195	9,890	136	45	126	131	262	176	3
d	Longitudinal	Room	8,270	88	33	151	154	239	187	2
d	,,	−195	10,800	160	56	158	164	318	220	1
d	Transverse	Room	7,870	89	27	97	100	186	127	1
d	,,	−195	9,500	126	57	130	134	256	191	2
e	Longitudinal	Room	8,020	78	27	128	131	206	158	1
e	,,	−195	9,740	115	39	155	160	270	199	1
e	Transverse	Room	6,560	Material failed at loading hole						
e	,,	−195	9,480	114	33	119	123	233	157	2
f	Longitudinal	Room	8,190	86	34	121	124	207	158	1
f	,,	−195	10,420	143	48	167	173	310	221	1
f	Transverse	Room	8,150	84	29	87	90	171	119	1
f	,,	−195	9,930	113	33	128	132	241	165	3

* Usually the gross values only are quoted (*see* ref. p. 512)

† Fracture appearance classifications:
1. Failure in a single oblique plane
2. Failure in two oblique planes
3. Failure normal to the plane of the material
4. Failure with a double cup appearance

10—Double cup fracture in longitudinal test piece type *f* of 50SWP extrusion. Tear commenced ahead of weld; tested at −195°C. ×3

value. The second correction took into account the work done in extending the loading holes in the test piece. Unslotted test pieces of each material were loaded incrementally, the load being removed and the loading hole diameters being measured between successive loadings. Graphs of load against plastic deformation were thus obtained (Fig. 5). Hence it was possible in each test, knowing the maximum load and the permanent extensions of the loading holes, to estimate the amount of work done in producing these extensions.

In their original paper Kahn and Imbembo did not make these corrections, and in comparing test results from different sources it is necessary to determine whether or not such corrections have been made.

RESULTS

Tensile tests

The tensile properties of welded and unwelded specimens are shown in Table II. All the materials showed a considerable increase in tensile strength if cooled to −195°C., without loss in elongation. The

Table III

Results of Navy tear tests on 50SWP (H.9) extruded section and pipe and on 54SM (N.5) extruded section

Type of Test Piece (see Fig. 1)	*Grain Direction*	*Test Temp., °C.*	*Maximum Load, lb*	*Work done on test piece,* ft.lb*						*Fracture appearance† (see Fig. 6)*
				To start tear		*To propagate tear*		*Total*		
				Gross	*Corrected*	*Gross*	*Corrected*	*Gross*	*Corrected*	
50SWP extrusion (HE.9.WP)										
a	Longitudinal	Room	8,600	56	34	149	153	205	187	4
a	,,	−195	11,130	93	52	166	173	259	225	3
a	Transverse	Room	9,000	69	38	122	125	191	163	1 and 2
a	,,	−195	11,180	86	40	136	144	222	184	1
b	Longitudinal	Room	6,600	53	29	114	117	167	146	1
b	,,	−195	7,540	42	26	80	82	122	108	3
b	Transverse	Room	6,550	67	29	121	123	188	152	1
b	,,	−195	7,200	53	27	71	74	124	101	3
c	Longitudinal	Room	6,050	79	41	121	123	200	164	4
c	,,	−195	8,450	109	48	196	200	305	248	4
c	Transverse	Room	6,100	54	27	128	130	182	157	4
c	,,	−195	8,750	108	44	193	197	301	241	3
d	Longitudinal	Room	7,040	170	58	161	163	331	221	4
d	,,	−195	9,950	285	79	199	203	484	282	4
d	Transverse	Room	6,390	110	41	120	123	230	164	1
d	,,	−195	8,070	149	55	159	162	308	217	3 and 4
e	Longitudinal	Room	5,850	43	6	172	174	215	180	4
e	,,	−195	7,900	94	19	211	215	305	234	4
e	Transverse	Room	5,610	48	7	173	175	221	182	1 and 4
e	,,	−195	7,170	65	8	226	229	291	237	1 and 2
f	Longitudinal	Room	7,535	143	57	159	162	302	219	4
f	,,	−195	10,270	215	94	201	206	416	300	4
f	Transverse	Room	6,840	103	24	164	167	267	191	4 and 1
f	,,	−195	10,250	215	67	209	214	424	281	1 and 2
50SWP pipe (HV.9.WP)										
a	Longitudinal	Room	8,780	58	32	125	129	183	161	1 and 2
a	,,	−195	10,630	58	28	123	128	181	156	3
a	Transverse	Room	8,660	50	31	75	78	125	109	1
a	,,	−195	10,610	54	28	90	95	144	123	1
a‡	Longitudinal	Room	8,710	65	32	117	120	183	152	1 and 2
a‡	,,	−195	10,540	75	41	129	134	204	175	1 and 2
a§	Transverse	Room	8,880	67	31	81	105	148	136	1
a§	,,	−195	10,730	54	23	88	93	142	116	2 and 3
b	Longitudinal	Room	6,000	50	27	137	138	187	165	1
b	,,	−195	7,750	48	29	96	99	144	128	3
c	Longitudinal	Room	5,820	56	32	130	132	186	164	4
c	,,	−195	8,500	92	53	213	216	305	269	4
d	Transverse	Room	7,700	167	57	98	101	265	158	1 and 2
d	,,	−195	9,950	177	56	190	194	367	250	3 and 2
e	Transverse	Room	5,670	60	15	142	143	202	158	3 and 4
e	,,	−195	8,050	75	9	183	186	258	195	2

* Usually the gross values only are quoted (*see* ref. p. 512)
‡ Tear began at and ran normal to porthole join
§ Tear began in and ran along porthole join

† Fracture appearance classifications:
1. Failure in a single oblique plane
2. Failure in two oblique planes
3. Failure normal to the plane of the material
4. Failure with a double cup appearance

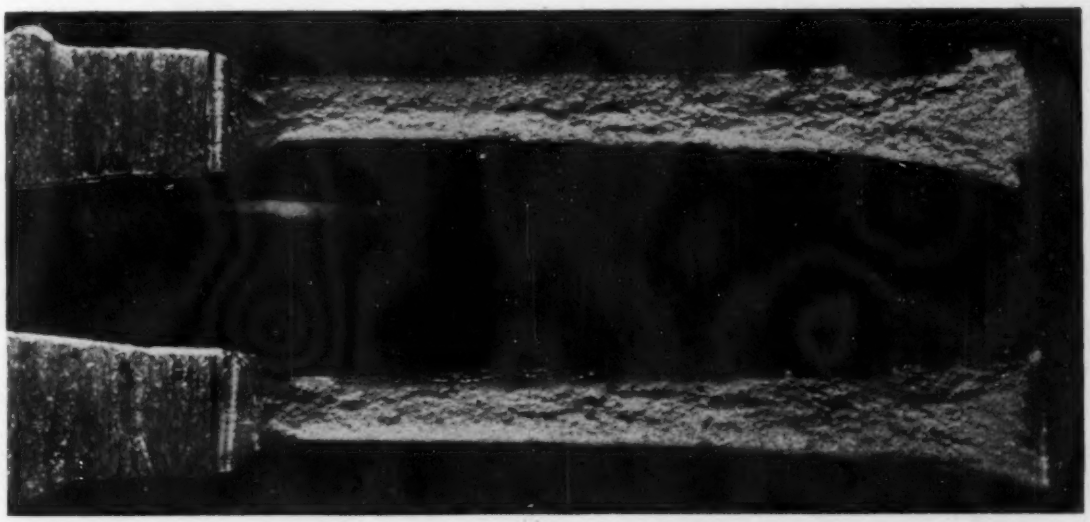
Table III (*continued*)

Results of Navy tear tests on 50SWP (H.9) extruded section and pipe and on 54SM (N.5) extruded section

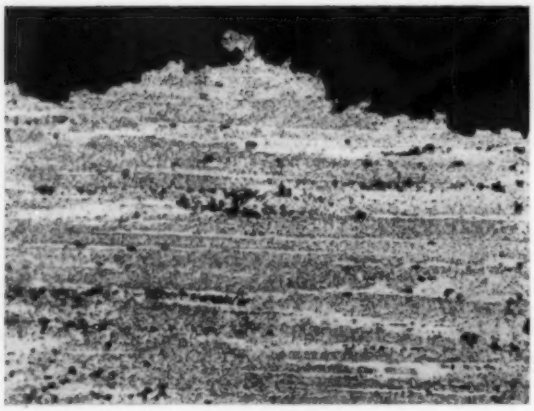
Type of Test Piece (*see Fig. 1*)	*Grain Direction*	*Test Temp., °C.*	*Maximum Load, lb*	*Work done on test piece,* ft.lb*						*Fracture appearance†* (*see Fig. 6*)
				To start tear		*To propagate tear*		*Total*		
				Gross	*Corrected*	*Gross*	*Corrected*	*Gross*	*Corrected*	
54SM extrusion (NE.5.M)										
a	Longitudinal	Room	9,400	91	46	148	156	239	202	1
a	,,	−195	11,870	149	62	156	138	305	200	1
a	Transverse	Room	9,330	91	44	92	96	183	140	1
a	,,	−195	11,350	127	46	117	120	244	166	3
b	Longitudinal	Room	8,150	83	23	128	131	211	154	1
b	,,	−195	9,850	111	28	95	100	206	128	3
b	Transverse	Room	8,350	132	30	118	121	250	151	1
b	,,	−195	9,975	141	33	105	119	246	152	3
c	Longitudinal	Room	7,820	79	32	121	124	200	156	1
c	,,	−195	10,200	139	46	164	169	303	215	1
c	Transverse	Room	7,810	97	37	99	102	196	139	1
c	,,	−195	9,890	136	45	126	131	262	176	3
d	Longitudinal	Room	8,270	88	33	151	154	239	187	2
d	,,	−195	10,800	160	56	158	164	318	220	1
d	Transverse	Room	7,870	89	27	97	100	186	127	1
d	,,	−195	9,500	126	57	130	134	256	191	2
e	Longitudinal	Room	8,020	78	27	128	131	206	158	1
e	,,	−195	9,740	115	39	155	160	270	199	1
e	Transverse	Room	6,560	Material failed at loading hole						
e	,,	−195	9,480	114	33	119	123	233	157	2
f	Longitudinal	Room	8,190	86	34	121	124	207	158	1
f	,,	−195	10,420	143	48	167	173	310	221	1
f	Transverse	Room	8,150	84	29	87	90	171	119	1
f	,,	−195	9,930	113	33	128	132	241	165	3

* Usually the gross values only are quoted (*see* ref. p. 512)

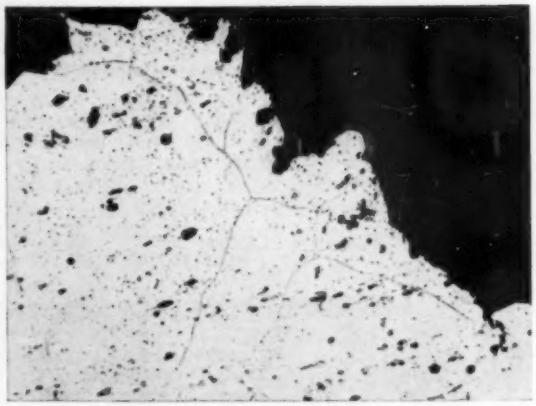
† Fracture appearance classifications:
1. Failure in a single oblique plane
2. Failure in two oblique planes
3. Failure normal to the plane of the material
4. Failure with a double cup appearance



10—Double cup fracture in longitudinal test piece type *f* of 50SWP extrusion. Tear commenced ahead of weld; tested at −195°C. ×3



11—Navy tear test fracture of specimen type *a* of 54SM extrusion; section in plane of material. Etched in dilute Tucker's reagent ×150



12—Navy tear test fracture of specimen type *a* of 50SWP extrusion; section in plane of material. Etched in dilute Tucker's reagent ×150

50SWP section after welding did not have an ultimate strength at room temperature as high as is usually attained, *i.e.*, 9 tons/sq.in. It is thought that the small sizes of the pieces welded together led to the whole of the material becoming heated so that the heat in the weld was conducted away too slowly for the full properties to be developed. When large pieces of 50S pipe material were welded together and small test coupons cut from them, ultimate stresses between 9½ and 10 tons/sq.in. were obtained.

Navy tear test

The results of Navy tear tests are given in Table III.

Satisfactorily high values of energy to propagate cracks were obtained in all tests and, in general, tests in liquid air gave higher values than those made at room temperature.

Relatively low corrected values of energy required to initiate tearing were obtained for the 50SWP type *e* specimens. On the other hand, both the 50SWP *b* type test pieces and the 54SM *e* type test pieces gave fairly high results. It is possible that the difference of behaviour is related to the variations of hardness across the specimens which inevitably occur on welding, and to differences observed in the shape of the load/extension diagrams. Both of these points need further study.

The appearance of the fracture surfaces could be grouped into four classes, shown diagrammatically in Fig. 6. The type of every fracture is given in Table III; failures in a single shearing plane and in a double cup form predominated. Flat fractures normal to the plane of the specimens usually occurred along a weld. An example of each type of fracture appearance is shown in Figs. 7–10.

No difference could be detected between the microstructures of the specimens tested at room temperature and those tested at −195°C. Likewise, microexamination revealed no differences corresponding to the classification of tear appearance obtained from visual observation. Sections through the fractures (unwelded material) showed differences related to the alloy compositions as follows:

(*a*) The 54S material had an unrecrystallized structure and the fractures appeared to be almost wholly transcrystalline (Fig. 11).

(*b*) The 50S material had a completely recrystallized structure, and though the nature of the fractures was somewhat variable they were always at least partially intercrystalline (Fig. 12).

(*c*) The extent of intercrystalline failure was most evident in the specimens of 10 in. dia. pipe.

(*d*) The failures along welds were very similar for the two alloys; a fine dendritic structure characterized the material in the welds.

CONCLUSIONS

One of the most disturbing aspects of brittle failure is the great rapidity with which a crack, once formed, extends, little or no additional energy being needed to cause this extension. Neither of the materials tested showed any tendency to exhibit this type of behaviour. Furthermore, neither of them suffered from a deterioration of tensile properties at the low temperature (−195°C.) compared with their behaviour at room temperature, nor had one alloy any appreciable advantage over the other in the extruded form in the $\frac{5}{16}$ in. thickness tested. The choice of thickness of the material used for testing was determined by the specific application envisaged but it is intended to carry out further work on thicker material.

Tear Tests on Aluminium–Magnesium Alloy Plate

By J. Sawkill, Ph.D., and D. James, M.A.

SYNOPSIS

TEAR tests have been carried out on one aluminium-magnesium alloy at room temperature and at −195°C., on plain and welded specimens. Most of the specimens were made from as-rolled ¾ in. plate, but a few were made from ⅜ in. plate. In addition, some Charpy and tensile tests were carried out.

A consideration of the energy values of the tear test shows that there is no serious embrittlement of the alloy, or enhanced notch sensitivity at low temperatures. For some specimens cracking in the plane of the plate occurs, and it is thought that this is caused by large tensile forces developed in the short transverse direction. This direction has been shown to have poorer properties than the rolling and long transverse directions. It is considered that the cracking is not a serious feature in the use of this alloy at low temperatures.

CERTAIN aluminium-magnesium alloys are being considered for structural use at sub-zero temperatures, and there has been some concern at a report that these alloys become notch sensitive at such temperatures. This account is concerned with the testing of one alloy, R.T.I. 224, produced by Reynolds T.I. Aluminium Ltd. This is a non heat-treatable alloy containing, nominally, 4·2% Mg and 0·45% Mn, and is produced as plate to BS 1477 NP 5/6–M.

The notch sensitivity has been evaluated by means of the Navy tear test, as described by Kahn and Imbembo.* The testing has been done at ambient temperature and at −195°C. (*i.e.*, in liquid nitrogen), on plain and welded specimens. Most of the work has been on ¾ in. thick plate, as rolled. A few tests have been made on ⅜ in. plate. In addition, some Charpy and tensile test have been carried out.

SPECIMENS

Tear test—The specimens measured 3 × 5 in. in the full plate thickness and were notched to provide for asymmetric loading in tension with an initial eccentricity of 0·039 in., as shown in Fig. 1. For welded specimens the axis of the weld was along the line *XX*. Four types of specimen were prepared:

(i) Plain, with tear to propagate in the rolling direction
(ii) Plain, with tear to propagate across the rolling direction
(iii) Welded, with tear to propagate in weld; weld in the rolling direction
(iv) Welded, with tear to propagate in weld; weld across the rolling direction.

Three specimens of each type were tested at both ambient temperature and at −195°C.

Charpy test—Standard notched Charpy specimens were machined from ¾ in. plate. Two types of specimen were prepared:

(i) To fracture in rolling direction
(ii) To fracture across rolling direction.

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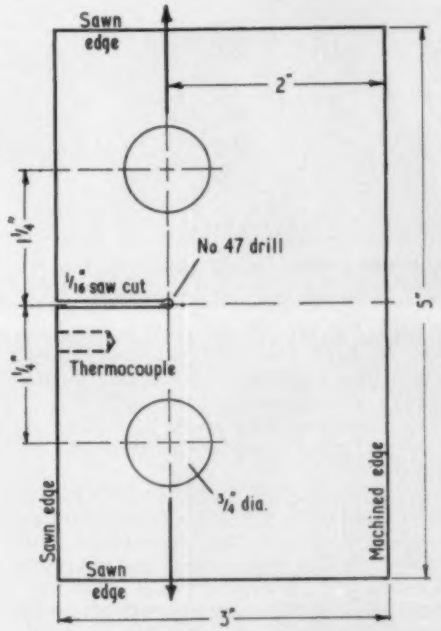


1—Tear test specimen

* N. A. KAHN and E. A. IMBEMBO: *Welding J.*, 1948, vol. 27, p. 169*s*; *Ibid*, 1949, vol. 28, p. 153*s*.

Tensile test—Very small tensile test specimens were machined from $\frac{3}{4}$ in. plate. The diameter was 0·178 in., giving a cross-sectional area of $\frac{1}{40}$ sq.in., and the length of the parallel portion was 0·38 in. Identical specimens were machined from three directions. These were the longitudinal (rolling) direction, and the two transverse directions.

Welding

The plate was welded at the Tube Investments Technological Centre, using the inert-gas metal-arc process, with the welding head moved mechanically. The shielding gas was commercial-grade argon supplied at a flow of 75 cu.ft/hr. The electrode wire, $\frac{3}{32}$ in. dia., was an NW6 alloy.

The edges of the plate were prepared to a 90° double **V** with a hand-ground $\frac{1}{4}$ in. root face, and were wire brushed and degreased by solvent wiping before welding.

Each plate was welded in two passes, one from each side, with the torch held vertically. The current was 200 amp with an arc voltage of 23V; the wire feed speed was 204 in./min and the welding speed 12 in./min.

All joints were radiographed after welding.

TESTING CONDITIONS

Tear tests—The tests were carried out on a Denison Universal testing machine, using the 25 ton load range. The specimen was pinned between steel cheek pieces, which in turn were pinned to a steel plate held in the wedge grips. For the sub-zero tests a metal can was built on to the lower steel plate and around the specimen. Liquid nitrogen was poured in until the temperature of the specimen, as recorded by a copper–constantan thermocouple, reached −195°C. At this stage the specimen was completely immersed in liquid nitrogen.

In the original work of Kahn and Imbembo a fixed crosshead speed of 0·1 in./min was used. Here, it was not possible to do this, and instead the load was applied at a constant rate, using a load pacer. The rate was full scale, *i.e.*, 25 tons, in 5 min. After the maximum load was reached no adjustments were made, the test looked after itself, and was completed in a few minutes.

The load/extension curve was obtained from an autographic recorder which measured a tenfold magnification of the crosshead movement. As the extension is not taken off the specimen, the true load/extension axes are not orthogonal, and a correction has to be made for the extension of all the gripping pieces. This correction can be made, and the true axes found, by substituting a 'rigid' link for the specimen. A suitable steel block was used to find the true axes.

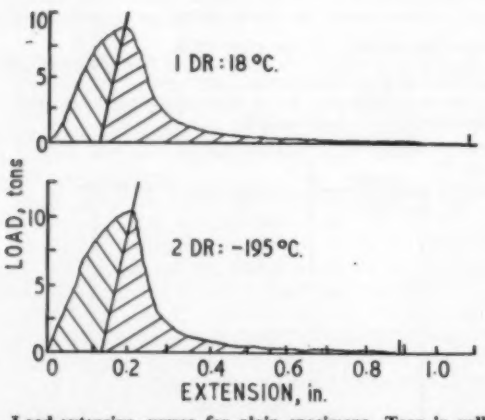
Charpy test—The impact testing was done on an Avery machine using the 120 ft.lb range. For sub-zero tests the specimens were immersed in liquid nitrogen for some time, and were then transferred, quickly, to the anvil. No temperature measurements were made.

Tensile test—Tensile testing was done on a Hounsfield tensometer. With such small specimens it was not possible to measure the elongation from a gauge length. Instead the elongation was estimated from a knowledge of the parallel portion and the overall increase in length.

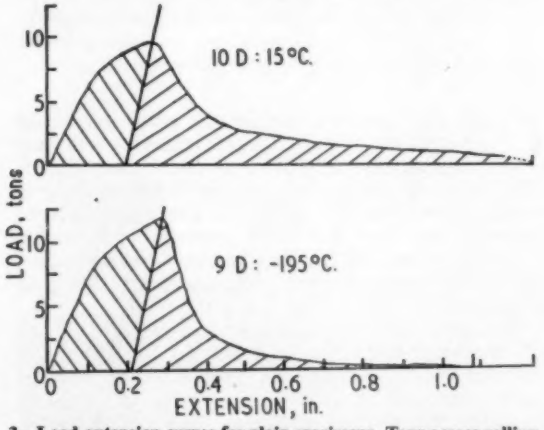
QUANTITATIVE EVALUATION

Typical load/extension curves, as taken from the autographic recorder, are shown in Fig. 2. It is assumed that in the tear test the load increases until a crack is formed, and it then decreases as the crack, or tear, is propagated. The area under the curve represents the work done on the specimen, and, with the previous assumption, this may be divided into two parts. The area up to the point of maximum load represents the energy input for crack initiation. The area after this point represents the additional energy input for crack propagation. These areas were determined from the record of the test, firstly by drawing a line through the maximum load point parallel to the true load axis, and secondly by measuring the areas, as indicated in Fig. 2, with a planimeter. The area values were then converted to energy values, in foot-pounds.

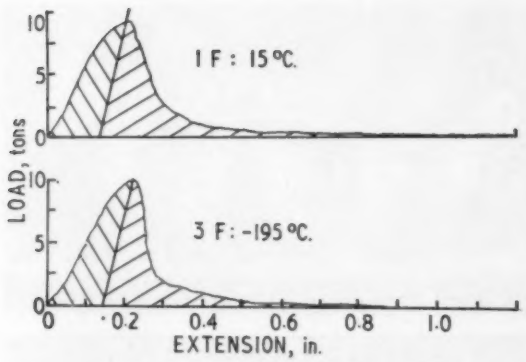
It must be stressed that these energy values represent energy input to the whole specimen and not those for the notched region alone. For example, in the first part of the curve, energy is used in deforming the specimen plastically, and as stored elastic energy. The former is illustrated by the non-circularity of the



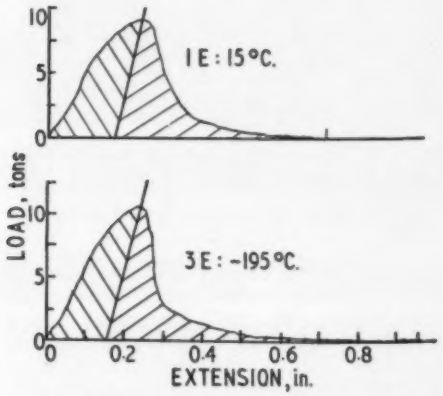
2—Load-extension curves for plain specimens. Tear in rolling direction



3—Load-extension curves for plain specimens. Tear across rolling direction



4—Load-extension curves for welded specimens. Tear in weld; weld in rolling direction



5—Load-extension curves for welded specimens. Tear in weld; weld across rolling direction

original ¾ in. dia. holes after the test. The latter is illustrated by loading and unloading. In unloading from near the maximum load point the curve does not follow the true load axis, but falls at a less steep angle. The area between these two lines gives the elastic energy stored at the point of unloading. Furthermore, this stored energy is used up in the propagation part of the process. From one experiment the elastic energy at maximum load was estimated at 28 ft.lb, but the value may be expected to vary from specimen to specimen.

For all of these reasons, it is considered best to give energy values only in terms of energy input to the specimen.

RESULTS

Tear tests on ¾ in. plate—The results are given with the following symbols:

I=energy input for crack initiation in ft.lb

P=additional energy input for propagation, in ft.lb

ML=maximum load, in tons

RT=room temperature, *ca* 15°C.

In Table I the results for the plain specimens are recorded, and Table II gives the results for welded specimens. In Figs. 2–5 examples are shown of the test

Table I

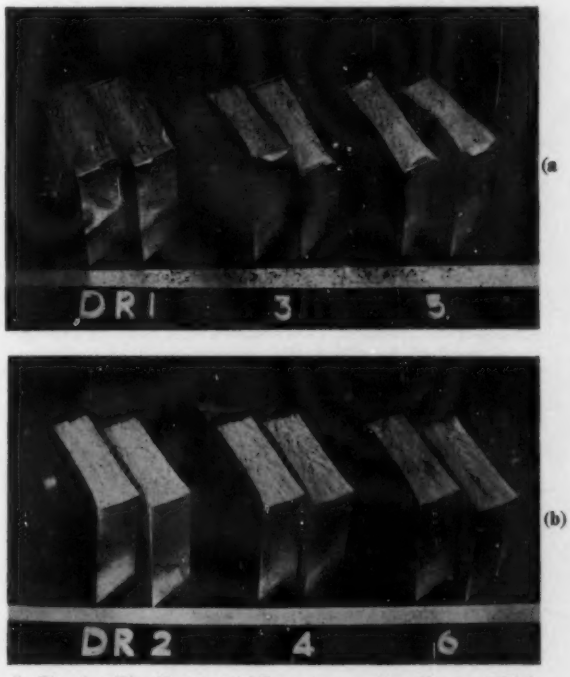
Plain specimens in ¾ in. plate

Temp., °C.	*Specimen*	*I*	*P*	*ML*
Tear in rolling direction				
RT	1DR	137	215	9·00
RT	3DR	154	208	9·05
RT	5DR	162	202	9·50
−195	2DR	202	215	11·25
−195	4DR	215	230	11·30
−195	6DR	204	208	11·50
Tear across rolling direction				
RT	7D	225	615	9·75
RT	10D	284	475	9·50
RT	13D	233	342	9·80
−195	9D	330	313	11·85
−195	12D	252	322	11·50
−195	15D	325	330	11·50

Table II

Welded specimens in ¾ in. plate

Temp., °C.	*Specimen*	*I*	*P*	*ML*
Tear in weld, weld in rolling direction				
RT	1F	141	239	9·10
RT	4F	161	247	8·80
RT	7F	118	222	8·15
−195	3F	158	180	10·00
−195	6F	147	184	9·65
−195	9F	154	160	10·55
Tear in weld, weld across rolling direction				
RT	1E	180	216	9·25
RT	4E	200	234	9·43
RT	7E	171	211	8·85
−195	3E	196	240	10·60
−195	6E	185	238	10·30
−195	9E	147	214	10·15



6—Tear in rolling direction: (*a*) Room temperature; (*b*) at −195°C.

Table III
Tensile tests

Specimen	*U.T.S., tons/sq.in.*	*Elong., %*	*Red. in area, %*
Longitudinal			
1L	20·2	29·0	47
2L	20·4	29·8	43
3L	20·3	29·0	40
Long-transverse			
1T	19·6	34·7	49
2T	18·6	34·0	40
3T	20·0	34·2	47
Short-transverse			
1ST	18·8	23·2	23
2ST	19·3	22·6	25
3ST	19·2	21·3	24

record for all types of test. Photographs of all specimens after fracture are shown in Figs. 6–9.

Charpy tests on ¾ in. plate—Five specimens were tested for each condition, and the mean results were:

Fracture in rolling direction:		
Energy absorbed, ft.lb	*RT*	15·2
	Low temperature	14·2
Fracture across rolling direction:		
Energy absorbed, ft.lb	*RT*	27·7
	Low temperature	20·7

Tensile tests—The results of the room temperature tests are set out in Table III.

DISCUSSION OF RESULTS

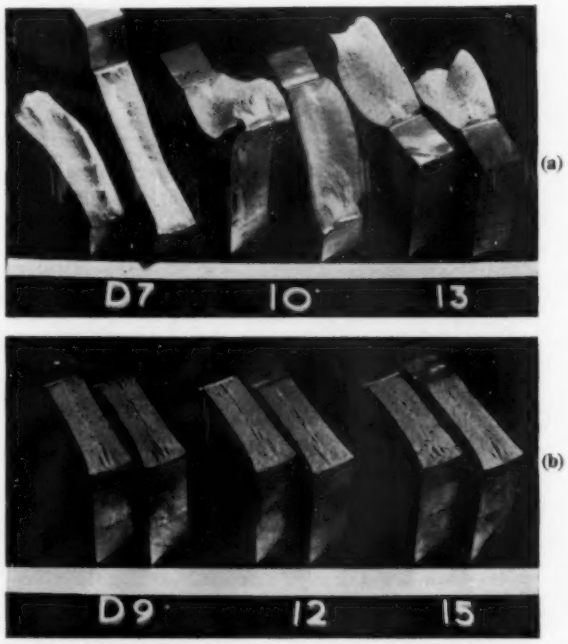
Plain specimens

The main conclusions to be drawn from Table I are:

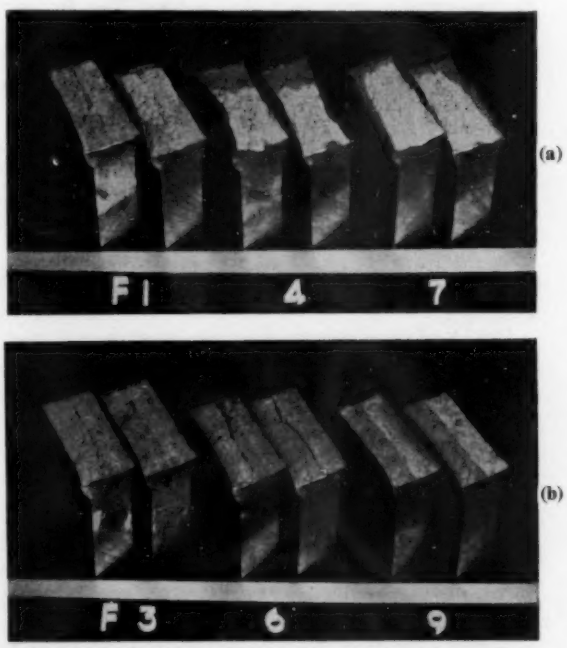
(i) The energy input for crack initiation is greater at −195°C. than at 15°C.
(ii) For tearing in the rolling direction there is no change in the energy for propagation with temperature
(iii) For tearing across the rolling direction, the energy for propagation decreases with temperature
(iv) The maximum load is always greater, by about 25%, at −195°C. than at room temperature
(v) Both the energy for initiation and for propagation are lower when the tear is in the rolling direction than when across it.

Kahn and Imbembo used the tear test primarily for following the ductile–brittle transition in steel plate. Whereas the energy for propagation was substantial in the ductile range, it dropped to almost zero in the brittle range. With the load/extension curves and the appearance of the fracture surface it was possible to follow the transition. It is evident that there is not the same type of ductile–brittle transition in the aluminium alloy plate. The lowest value recorded for the energy for propagation is 202 ft.lb, compared with zero for steel plate.

None of the five conclusions listed should cause concern; in fact, some of them indicate better properties at −195°C. than at room temperature. Thus, the load for crack initiation, and the energy for initiation, are greater at the lower temperature. The variation of energy values with direction might have been predicted, particularly the energy input for propagation. A metallographic examination showed that the grains were elongated in the direction of rolling, as were the intermetallic constituents. A tear across the rolling direction would have more grain boundaries and inter-metallic constituents to surmount than one in the rolling direction, and so the energy for propagation would be greater than for a tear along the rolling direction. The Charpy results show the variation with direction also, and indicate that the metal is less ductile at low temperature.



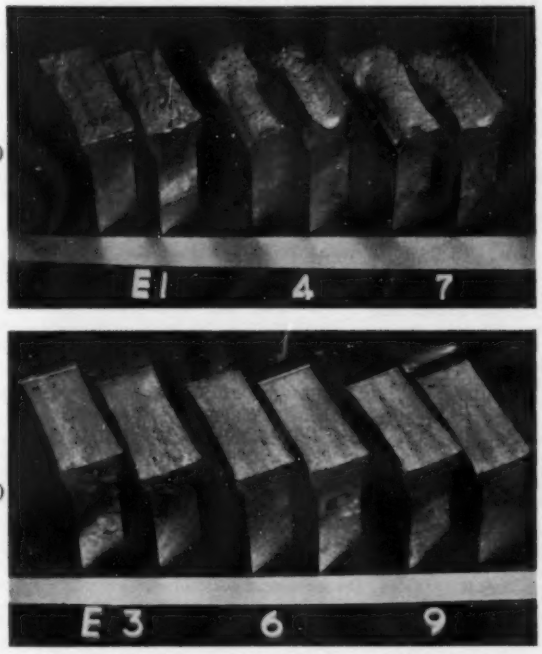
7—Tear across rolling direction. (*a*) Room temperature; (*b*) at −195°C.



8—Tear in weld. Weld in rolling direction. (*a*) Room temperature; (*b*) at −195°C.

The energy values of the tear test are easy to interpret when a metal becomes brittle, but it is difficult to attach a physical significance to small changes in what is essentially ductile behaviour. The examination of the fracture surface is an essential part of the test. Although the changes in energy values are difficult to interpret precisely, it is obvious from the photographs of the specimens that the metal is less ductile at low temperatures. At such temperatures there is a cleaner, flatter, fracture, although there is still a good deal of 'necking'.

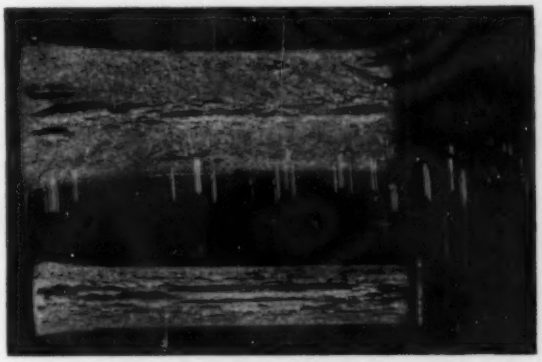
One feature, however, does give rise to some concern, and this is the cracking found in the plane of the plate in some specimens. It is noticeable in Fig. 7*b* in particular. Figure 10 shows the phenomenon in greater detail in $\frac{3}{4}$ in. and $\frac{3}{8}$ in. plate, and Fig. 11 shows it in Charpy test pieces. Attempts were made to relate this effect to the microstructure. Sections were taken



9—Tear in weld. Weld across rolling direction. (*a*) Room temperature; (*b*) at −195°C.

through the cracks, but it is not possible to state that the cracking was definitely associated with grain boundaries or intermetallic constituents.

It is thought that the cracking is caused by large tensile forces developed during the test in the short transverse direction, together with poor mechanical properties in this direction. The tensile tests, as recorded in Table III, were made to determine the latter point. The elongation values are higher here than would be measured on a standard tensile piece, because the deformation of the neck is a greater percentage of the total deformation with a small test piece. Nevertheless, the results indicate that at room temperature the short transverse direction exhibits lower elongation, lower reduction in area, and lower strength, than the other directions. The cracking is seen to be



10—Cracking in $\frac{3}{4}$ in. and $\frac{3}{8}$ in. plate

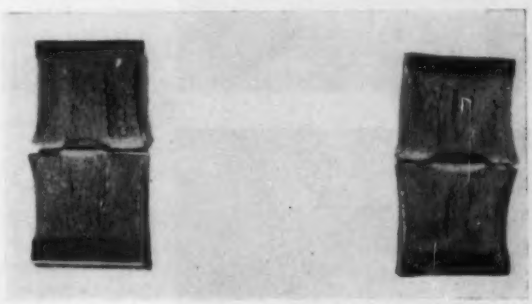
more severe in the $\frac{3}{8}$ in. plate than in the $\frac{3}{4}$ in. material, which may mean that grain shape is an important factor. Certainly, the thickness of the plate, in itself, is not important. Specimens $\frac{3}{8}$ in. thick were made from $\frac{3}{4}$ in. plate, and showed much less severe cracking than the as-rolled $\frac{3}{8}$ in. plate.

The cracking usually does not occur at room temperature and is worst in the $\frac{3}{4}$ in. plate when the tear is across the rolling direction and at −195°C. This was confirmed for the $\frac{3}{8}$ in. plate. In this condition the maximum load is greatest, the energy for initiation greatest, and the energy for propagation large also. Thus one might expect the tensile forces in the short transverse direction also to be greatest in this test.

The phenomenon must be regarded as an indication of brittleness, but brittleness in one direction only. Figure 10 shows that as a whole the specimen is ductile, as indicated by the substantial reduction in thickness of the plate at the tear. This, together with the energy values, would indicate that the alloy can be safely used at low temperatures.

Welded specimens

The results for these specimens are similar to those for plain specimens, so far as the pattern of energy values is concerned. There is not so large a dependence on direction, and this is to be expected since it is mainly the weld metal that is being tested. The maximum load values are again higher at low temperature. The lowest value of the energy for propagation recorded was 160 ft.lb. There was no sign of the cracking phenomenon, which again might be expected, since it is not the plate as such that is being tested, but the weld. The lowest values for the energy for propagation



11—Cracking in Charpy test pieces

occur when the weld is in the rolling direction, and at −195°C. There is no explanation for the values being substantially lower than those for the remaining tests. Overall, the properties of the welds at low temperatures should not cause concern.

CONCLUSIONS

A consideration of the energy values in the tear test indicates that the particular alloy used in these tests is not seriously embrittled or notch sensitive at low temperatures, in either the plain or welded condition. In fact, from a structural viewpoint some properties are improved; *e.g.*, a greater load is needed for crack or tear initiation at low temperature, and the energy for initiation is also greater.

Inspection of the fracture surface shows some instances of cracking in the plane of the plate, which occurs mainly when the tear is across the rolling direction and at −195°C. This is when the maximum load, and the energy for initiation, are at their greatest values, and a high propagation energy obtains. It is thought that the cracking is caused by the tensile forces set up in the short transverse direction, a direction which has been shown to have the poorest properties. The phenomenon is not thought to be serious in this alloy, for there is still a fair measure of ductility, as indicated by the energy values and the contraction at the tear. The phenomenon might be more serious in those wrought alloys that have a greater variation of properties with direction, and, in particular, poor short transverse properties.

ACKNOWLEDGMENTS

This work was done in collaboration with Reynolds T.I. Aluminium Ltd., and Mr. P. Kyne is thanked for his helpful discussion and advice. The testing was carried out by Mr. A. C. Nunn, and the B.W.R.A. kindly assisted in the early stages of the work.

EXAMPLES OF WELDED CONSTRUCTION

Improved highway bridges

The James F. Lincoln Arc Welding Foundation recently announced the award of a highway bridge design competition. The award designs showed substantial savings in steel and cost, and also indicated that welded bridges were more beautiful and easier to maintain than riveted structures. Designs were judged on a basis of economical use of labour and steel, appearance, cost, general quality of detail, and other advantages of welded construction. One award winner showed that a four-span haunched girder bridge of welded construction took 24% less steel than if it had been made by conventional methods. Similarly an all welded arch bridge was completed, due to the ease of welded fabrication, in just over three months.

The Award Jury stated that the high quality of the designs, the excellent detail, and the size of the projects involved showed a maturity of welded design practice.

Aluminium hatch covers

In 1954 the first riveted aluminium MacGregor hatch covers were fitted experimentally in the ore carrier *Sunbrayton*, built by the Burntisland Shipbuilding Company for Saguenay Terminals Ltd.

Until the comparatively recent introduction general use of the inert-gas consumable electrode welding process, the development of aluminium covers was delayed by the impossibility of designing riveted structures to compete economically with welded steel structures.

The announcement that complete sets of welded aluminium MacGregor insulated deck covers are to be fitted in three Royal Mail Line ships, now being built at Belfast by Harland & Wolff Ltd., comes as a result of much development work and successful service experienced with existing installations. The design of these covers has been developed by MacGregor & Company Ltd. in conjunction with Saunders-Roe (Anglesey) Ltd., who are fabricating them from material supplied by the Northern Aluminium Company Ltd.

The weight saving by using aluminium has a dual advantage, for it allows the cargo deadweight to be increased and allows easier handling of the covers. The corrosion resistance and consequent reduction in maintenance can be accounted for, and all these factors outweigh the increased initial cost.

Dock gate for Queen Elizabeth Dock at Falmouth

The dock gate for the recently opened Queen Elizabeth Dock at Falmouth is largely of welded construction and with a weight of 500 tons it is probably the largest so far constructed. It was designed and manufactured by Sir William Arrol & Co. Ltd. and the site welding was done by the Lanarkshire Welding Co. Ltd. using mostly Rockweld Flexend and Vertend electrodes.

The gate is a box-type construction in one piece and was positioned by floating it and then flooding various compartments to bring it into an upright position.

1st Award

Haunched twin-span girder bridges at Houston, Texas

2nd Award

Box girder arch bridge on Dallas–Fort Worth Turnpike

Welded aluminium hatch covers

Low Temperature Properties of Aluminium–Magnesium Alloys

By R. E. Lismer, A.Met., F.I.M.

SYNOPSIS

INVESTIGATIONS into the behaviour of certain metals and alloys at low temperatures, down to that of liquid nitrogen, are being undertaken at the Research Laboratory of Lloyd's Register of Shipping. This paper reports the results of a series of tests made on certain light alloys in the wrought and welded conditions. Differences in the appearance of fractures, and the energy values to produce fracture, have been noted between alloys of varying magnesium content in the hot rolled, annealed or extruded conditions, and for some variation of weld quality.

Introduction

ABOUT two years ago, proposals were made to Lloyd's Register of Shipping for the use of aluminium-magnesium alloy plate as a material suitable for the construction of tanks for the marine transportation of liquified gases at temperatures down to −160°C. At that time there was only limited data available on the behaviour of these alloys at sub-zero temperatures, and it was considered that a series of tests should be carried out to determine the notch sensitivity and fracture characteristics at very low temperatures.

The tests reported in this paper were undertaken by the Research Department of that Society. The Aluminium Development Association co-operated in all the tests which were made on alloys of British manufacture.

It was initially decided to use specimens conforming to the U.S. Navy tear test and investigate the tearing properties of the proposed materials at low temperatures. This test had been widely applied to ferritic steels and has been detailed by Kahn and Imbembo.[1, 2]

The advantages of this test are:

(*a*) The tests could be conducted on the full plate thickness which had been proposed

(*b*) The depth of the test section behind the notch was suitable to determine any change in the mode of fracture that might occur between tests conducted at ambient and low temperatures

(*c*) The fracture would be of the tearing type, which would probably occur in the event of a failure in a large, low-temperature, storage vessel made from plates of the alloys under consideration. Some indication of the ductility that accompanied this type of fracture could therefore be obtained

(*d*) The load/extension curve could be drawn from measurements taken during each test. An indication of the energies to initiate a fracture or propagate a fracture could be obtained from each of these diagrams

(*e*) The tests could be applied to butt welds in the plates.

As a check on the fracture characteristics, it was decided to undertake a parallel series of tests of notched tensile type. The Tipper test[3, 4] was selected for this purpose. The same loading and low temperature apparatus were used to test the Navy tear and Tipper specimens.

DESCRIPTION OF APPARATUS

It was necessary to conduct the above types of tests within the range +16° to −196°C. Figure 1 shows the apparatus, which was adapted for use in the 50-ton Avery Universal Testing Machine. The load was applied to the specimen through high-tensile tension rods and pin-type shackles. The specimens were retained to the shackles by ¾ in. dia. pins, which were a sliding fit in the holes in the legs of the shackles. Whilst this type of attachment was normal for Navy tear test specimens, it was necessary to drill suitable holes in the 'gripping' sections of the Tipper test specimens. In some instances the width of these sections had to be increased for strength purposes.

The specimen and shackles were surrounded by two concentric containers. The inner aluminium container was 4½ in. dia. and was designed to hold special refrigerants or liquid nitrogen. The outer brass container for liquid nitrogen was 6½ in. dia. A brass liquid-tight locknut and gland were attached to the base of the inner container to form a seal around the shank of the lower tension rod. The lock-nut was surrounded by a low-power heating coil, which served to keep the gland free from ice. The container could be lowered for the purpose of changing the test specimens. The two containers were heavily insulated with 'Stillite' semi-rigid Thermobloc slabs, which were fitted in the space between the outer brass container and an additional outer brass cylinder.

For measuring the temperature of the specimen, 36 s.w.g. enamelled copper–constantan thermocouples were used in conjunction with a quick reading potentiometer. Three couples were used, one adjacent to the root of the notch in the specimen, whilst others were positioned at the top and bottom of the specimen.

Calibration

The deformation of the test specimen, in the line of the load, was measured by the movement of the beam of the testing machine relative to a datum. During a test, this movement was recorded by a dial gauge reading to 0·001 in. The dial gauge reading represented the

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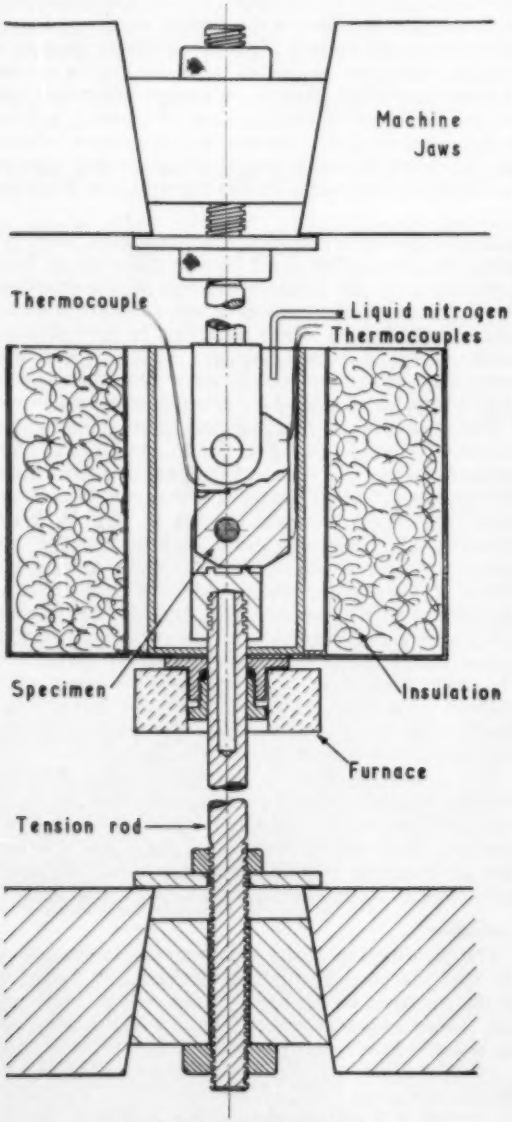
Mr. Lismer is a member of the Research Staff of Lloyd's Register of Shipping, Crawley.

Table I
Specifications and analyses of alloys

Specification No.	*Material*	*Analysis, %* *Mg*	*Mn*	*Si*	*Fe*	*Cu*	*Zn*	*Cr*	*Ti*
BS. 1477 NP.5/6		3·5/5·5	1·0 max.	0·6 max.	0·75 max.	0·1 max.	0·1 max	0·5 max.	0·2 max.
	Alloy *A*	3·82	0·24	0·14	0·28	0·02	0·01	—	0·03
	Alloy *B*	4·8	0·33	0·15	0·28	0·05	0·06	—	0·03
	Alloy *C*	4·91	0·25	0·16	0·28	0·02	—	—	—
BS. 1476 NE.5		3·0/4·0	1·0 max.	0·6 max.	0·75 max.	0·1 max.	0·1 max.	0·5 max.	0·2 max.
	NE.5.M	3·34	0·57	0·22	0·30	0·029	—	—	0·01
U.S. Designation 5083		4·0/4·9	0·5/1·0	0·40 max.	0·40 max.	0·10 max.	0·25 max.	0·25 max.	0·15 max.
	5083	4·7	0·66	0·14	0·19	0·02	0·03	0·11	0·03
U.S. Designation 5356		4·5/5·5	0·05/0·20	Si+Fe=0·50 max.		0·10 max.	0·10 max.	0·05/0·20	0·06/0·20
	5356	4·8	0·12	0·10	0·20	0·05	0·01	0·098	0·15
BS. 1476 HE.9		0·4/0·9	—	0·3/0·7	0·60 max.	0·15 max.	—	—	—
	HE.9.WP	0·65	0·035	0·47	0·34	0·028	—	—	0·01



1—Diagram of testing apparatus

general deformation of the specimen between the inside edges of the pin holes, the distortion of the pin holes, and elastic and other movements in the associated loading equipment and measuring attachments.

A wide rectangular steel plate, $\frac{3}{4}$ in. thick and having holes positioned so that the shackles were close together, was loaded several times in the equipment to a load well above the maximum required for the tests on the aluminium alloy specimens. The steel section was used to calibrate the equipment up to a load of about 2 tons above that required. A series of calibrations indicated that it was reasonable to accept the elastic and other movements in the loading equipment and measuring attachments as systematic.

For each type and thickness of aluminium plate, a Navy tear test specimen was loaded in increments of 0·5 tons up to the maximum load. The specimen was removed from the testing machine after each loading and the deformation of the holes in line with the applied load was measured. This provided a measure of the deformation at the holes with load, and it was noted that very little plastic deformation occurred around the position of the notch until the later stages of the tests. As it was impractical to undertake such tests at low temperatures and to cover differences in properties between specimens from the same plate, the maximum load and final deformation of the holes were recorded in each test and a simple correction was applied in proportion to these quantities.

For convenience, the point of maximum load was used to represent the end of 'fracture initiation' and the beginning of the 'propagation of the fracture'. Errors involved in the method of allowing for the deformation of the holes therefore influenced the 'load increasing' section of the test. At room temperature, the notches could be observed during test and it was noted that cracks formed in these regions before the maximum load was attained. Although these initial cracks were relatively shallow, the depth appeared to vary in some tests on the same material. This influenced the computed 'energy for crack initiation' and it appeared that the two stages of the test could be more accurately expressed as the 'initiation and slow growth of cracks' and 'tearing of the material'. The former terminology has been retained for uniformity.

MATERIAL TESTED

With one exception, the aluminium alloys investigated were of the non-heat-treatable type:

(*a*) Alloys *A*, *B*, and *C* to BS. 1477, NP.5/6. Alloys *A* and *B* were supplied as hot rolled $\frac{3}{4}$ in. thick plate material, having Mg contents towards the bottom and top of the specified range, respectively. Alloy *C*, hot rolled 1 in. thick plate material, was of similar composition to that of alloy *B*

(*b*) Alloy NE.5.M to BS. 1476, supplied as extruded $\frac{3}{4}$ in. thick flat bar, 5 in wide

(*c*) Two alloys, conforming to the U.S. Aluminium Association Designations 5083 and 5356, of $\frac{3}{4}$ and $\frac{5}{16}$ in. plate thickness respectively

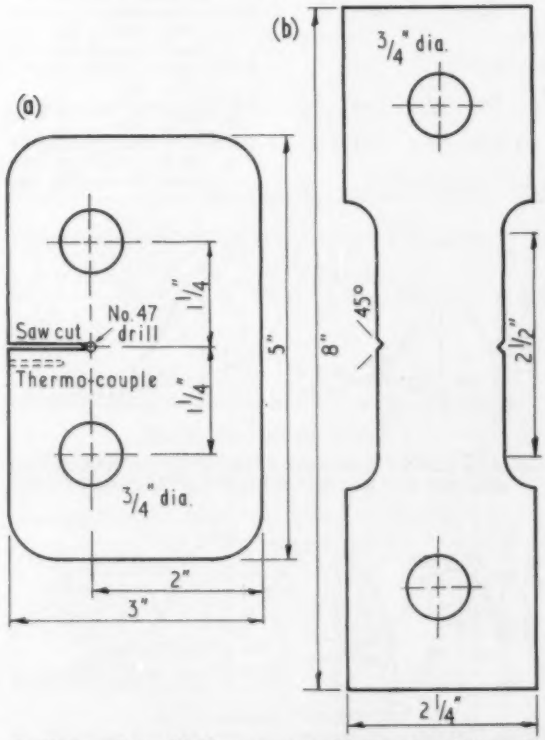
(*d*) Alloy HE.9.WP to BS. 1476, supplied as extruded $\frac{3}{4}$ in. thick flat bar, 5 in. wide and in the solution and precipitation treated condition.

The limits of composition allowed by the specifications, and the results of chemical analysis of all the materials tested, are given in Table I.

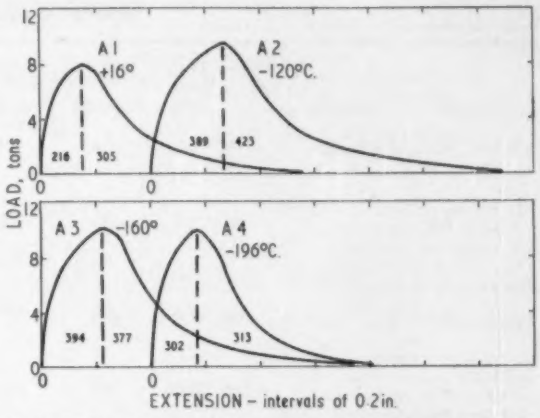
TESTING PROCEDURE

With the exception of the 1 in. thick plate alloy *C* which was reduced to $\frac{3}{4}$ in. thickness, all specimens for tests were of full plate thickness. The dimensions of the Navy tear test and Tipper notched tensile test specimens as used in this investigation, are shown in Fig. 2.

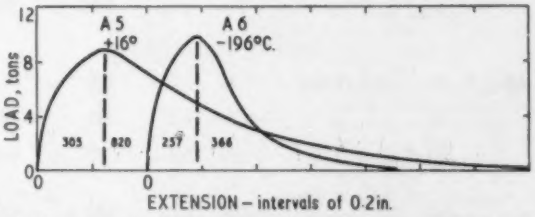
All tests were carried out under reasonably isothermal conditions and the temperatures were stabilized before commencing to load the specimen. Between 0° and −100°C. the specimens were immersed in petroleum ether; for −120° to −160°C., the refrigerant 'Arcton 6' was used. In both examples, cooling was obtained by introducing liquid nitrogen under low pressure into the annular space between the inner containers. For tests at −196°C., the innermost container was filled with liquid nitrogen.



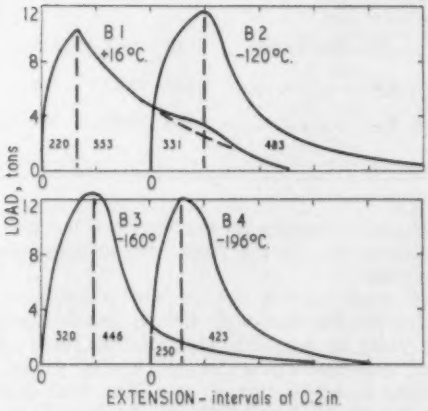
2—Test specimen: (*a*) Navy tear test; (*b*) Tipper notched tensile test



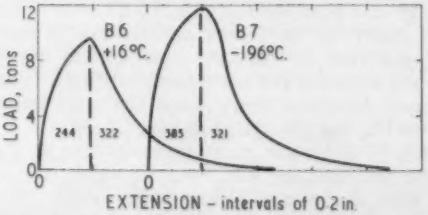
3—Navy tear test load/extension curves for hot rolled NP.5/6 alloy plate 0·735 in. thick, with 3·8% Mg



4—Navy tear test load/extension curves for annealed NP.5/6 alloy plate 0·735 in. thick, with 3·8% Mg



5—Navy tear test load/extension curves for hot rolled NP.5/6 alloy plate 0·75 in. thick, with 4·8% Mg



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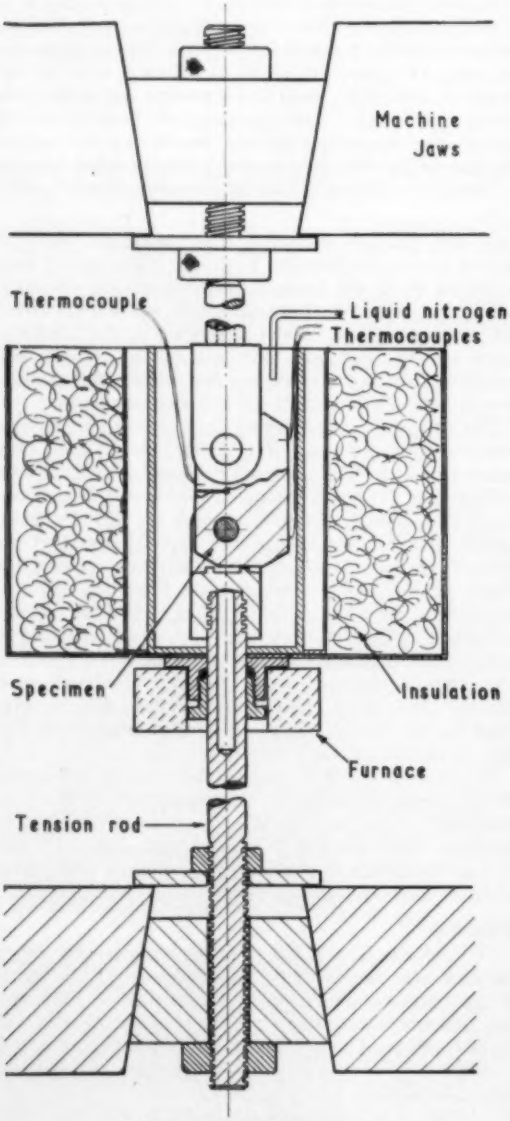

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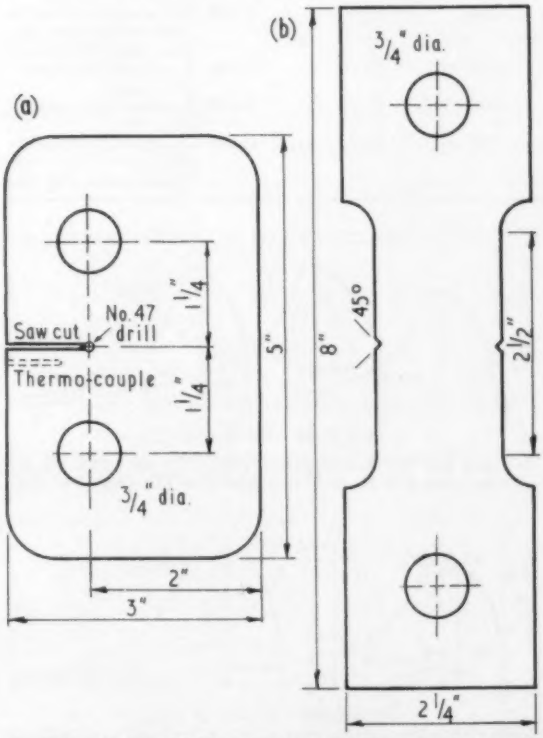
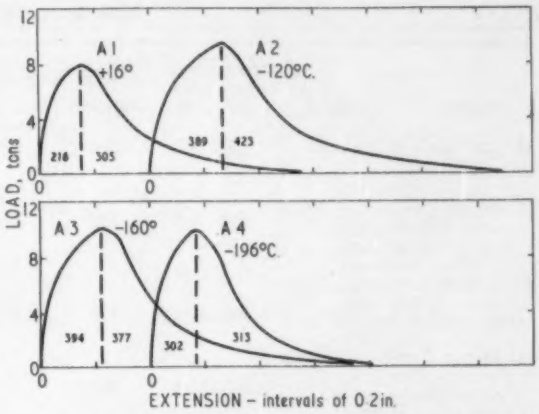
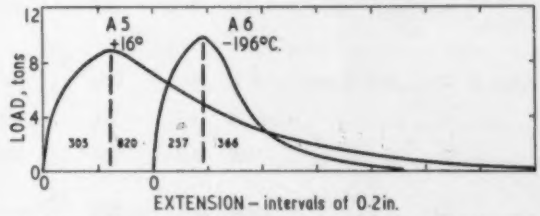


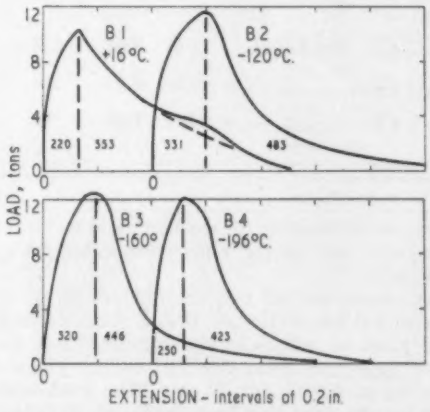
2—Test specimen: (*a*) Navy tear test; (*b*) Tipper notched tensile test



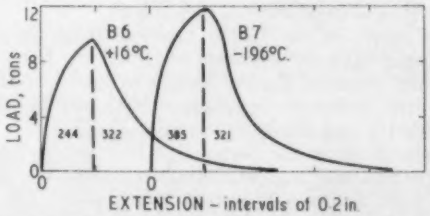
3—Navy tear test load/extension curves for hot rolled NP.5/6 alloy plate 0·735 in. thick, with 3·8% Mg



4—Navy tear test load/extension curves for annealed NP.5/6 alloy plate 0·735 in. thick, with 3·8% Mg



5—Navy tear test load/extension curves for hot rolled NP.5/6 alloy plate 0·75 in. thick, with 4·8% Mg



6—Navy tear test load/extension curves for annealed NP.5/6 alloy plate 0·75 in. thick, with 4·8% Mg

Table II—Results of Navy tear tests

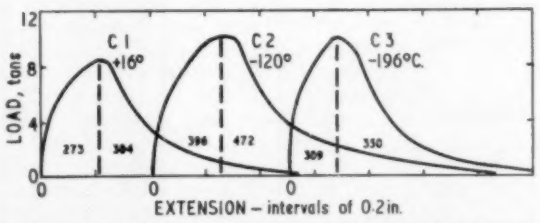
Material		Condition	Temp., of Test, °C.	Max. Load, tons	Corrected Energy Values, ft.lb Initiation	Propagation	Total	Ratio: Energy to propagate / Total energy %	Reduced Plate Thickness, in.	Appearance of Fractures
NP.5/6	A1	Hot Rolled	+16	8·2	216	305	521	58·5	0·640	Single oblique shear with thin mid-zone normal to plate surface (Fig. 12*a*)
	A2	„ „	−120	9·5	389	423	812	52·1	0·660	Single to double oblique shear with larger mid-zone normal to plate surface (Fig. 12*b*)
	A3	„ „	−160	10·0	394	377	771	48·9	0·590	Mainly normal to plate surface and a marked chevron pattern. (Fig. 12*c*)
	A4	„ „	−196	10·05	302	313	615	50·9	0·620	Completely normal to plate surface and a marked chevron pattern (Fig. 12*d*)
	A5	Annealed 400°C. for 20 min.	+16	8·95	305	820	1125	72·9	0·610	Double to single oblique shear (Fig.13*a*)
	A6	„ „	−196	9·65	257	366	623	58·8	0·590	Completely normal to plate surface and a marked chevron pattern (Fig. 13*b*)
Alloy *B*	B1	Hot Rolled	+16	10·25	220	553	773	71·6	0·670	Double oblique shear (Fig. 14*a*)
	B2	„ „	−120	11·5	331	483	814	59·3	0·640	Double oblique shear (Fig. 14*b*)
	B3	„ „	−160	12·3	320	446	766	58·2	0·670	Completely normal to plate surface and severely fissured (Fig.14*c*)
	B4	„ „	−196	12·15	250	423	673	62·9	0·670	As B3 (Fig. 14*d*)
	B5	„ „	−196	12·12	—	—	—	—	—	As B3
	B6	Annealed 400°C. for 20 min	+16	9·7	244	322	566	56·8	0·680	Single to double oblique (Fig. 15*a*)
	B7	„ „ „	−196	11·85	385	321	706	45·5	0·640	Completely normal to plate surface and severely fissured (Fig.15*b*)
Alloy *C*	C1	Hot Rolled	+16	8·45	273	304	577	52·7	0·700	Single oblique shear (Fig. 16*a*)
	C2	„ „	−120	10·0	396	472	868	54·4	0·655	Single oblique shear (Fig. 16*b*)
	C3	„ „	−196	10·05	309	350	659	53·1	0·670	Completely normal to plate surface and severely fissured (Fig. 16*c*)

The rate of straining for the Navy tear test was set at 0·1 in./min, and for the Tipper test at approximately 0·5 in./min.

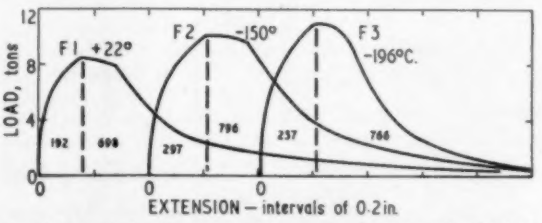
Tests were carried out on hot rolled plate and extruded flat bar section to British specifications, hot rolled plate to American designations, and on two British and one American hot rolled plates after annealing at 400°C. for 20 min. The load/extension curves for the Navy tear tests are shown in Figs. 3–11. The results of the tests, energy values computed by planimeter measurement of the areas under the curves, and the appearances of the fractures, are contained in Table II, and illustrated in Figs. 12–21.

The results of the Tipper notched tensile tests and the appearance of fractures obtained on hot rolled plate and extruded flat bar section to British specifications and American designation 5083, are contained in Table III, and illustrated in Figs. 22–27,

Table IV gives the percentage increase in load to fracture the tear test and notched tensile test specimens with decrease of testing temperature. The percentage reduction in plate thickness at fracture, as measured at a position about 0·5 in. from the root of



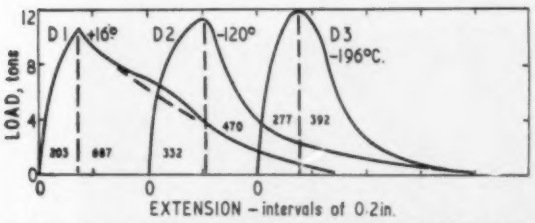
7—Navy tear test load/extension curves for hot rolled NP.5/6 alloy plate 0·75 in. thick machined from 1 in., with 4·9% Mg



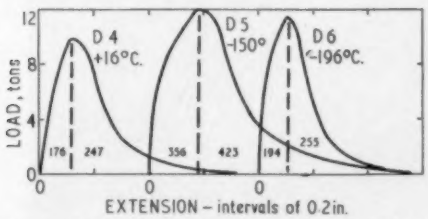
8—Navy tear test load/extension curves for extruded NE.5.M alloy plate 0·75 in. thick, with 3·34% Mg

Table II (*continued*)—Results of Navy tear tests

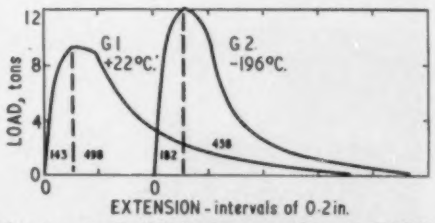
Material		Condition	Temp., of Test, °C.	Max. Load, tons	Corrected Energy Values, ft.lb Initiation	Propagation	Total	Ratio: Energy to propagate / Total energy %	Reduced Plate Thickness, in.	Appearance of Fractures
NE.5.M	F1	Extruded	+22	8·5	192	698	890	78·5	0·660	Single oblique (Fig. 17*a*)
	F2	„	−150	10·05	297	796	1093	72·8	0·630	Single oblique shear (Fig. 17*b*)
	F3	„	−196	9·9	257	766	1023	74·9	0·590	Completely normal to plate surface with fairly pronounced fissures (Fig. 17*c*)
5083	D1	Hot Rolled	+16	10·5	203	687	890	77·2	0·670	Double to single shear (Fig. 18*a*)
	D2	„ „	−120	11·25	332	470	802	58·6	0·655	Double oblique shear (Fig. 18*b*)
	D3	„ „	−196	12·0	277	392	669	58·6	0·635	Completely normal to plate surface, a marked chevron pattern with some fissures towards origin of fracture (Fig. 18*c*)
5083	D4	Annealed	+16	9·95	176	247	423	58·4	0·675	Mainly double oblique with mid-zone normal to plate surface (Fig. 19*a*)
	D5	„	−150	12·0	356	423	779	54·3	0·645	Mainly normal (Fig. 19*b*)
	D6	„	−196	11·6	194	255	449	56·8	0·680	Completely normal to plate surface with marked chevron pattern (Fig. 19*c*)
5356	E1	Hot Rolled	+16	6·2	187	229	416	55·0	0·440	Single oblique shear (Fig. 20*a*)
	E2	„ „	−120	6·7	246	300	546	55·0	0·420	Single oblique shear with fine fissures within mid-zone (Fig. 20*b*)
	E3	„ „	−150	6·7	229	317	546	58·0	0·400	Single oblique shear with further central fissuring (Fig. 20*c*)
	E4	„ „	−196	6·75	152	255	407	62·6	0·410	Completely normal to plate surface with pronounced fissures (Fig. 20*d*)
HE.9.WP	G1	Extruded	+22	9·2	143	498	641	77·7	0·650	Single oblique shear (Fig. 21*a*)
	G2	„	−196	12·05	182	458	640	71·6	0·670	Single oblique shear (Fig. 21*b*)



9—Navy tear test load/extension curves for hot rolled 5083 alloy plate 0·75 in. thick, with 4·7% Mg



10—Navy tear test load/extension curves for annealed 5083 alloy plate 0·75 in. thick, with 4·7% Mg



11—Navy tear test load/extension curves for extruded HE.9.WP alloy plate 0·75 in. thick, with 0·65% Mg

the notch for Navy tear test specimens and at the mid-width for notched tensile test specimens, are also included in Table IV. The agreement between comparable results for both types of tests is considered to be reasonable.

DISCUSSION OF RESULTS

Navy tear tests

(1) The results of the Navy tear tests are summarized graphically in Figs. 28 and 29. There was a general indication of increasing strength and ductility

Table III—Results of Tipper notched tensile tests

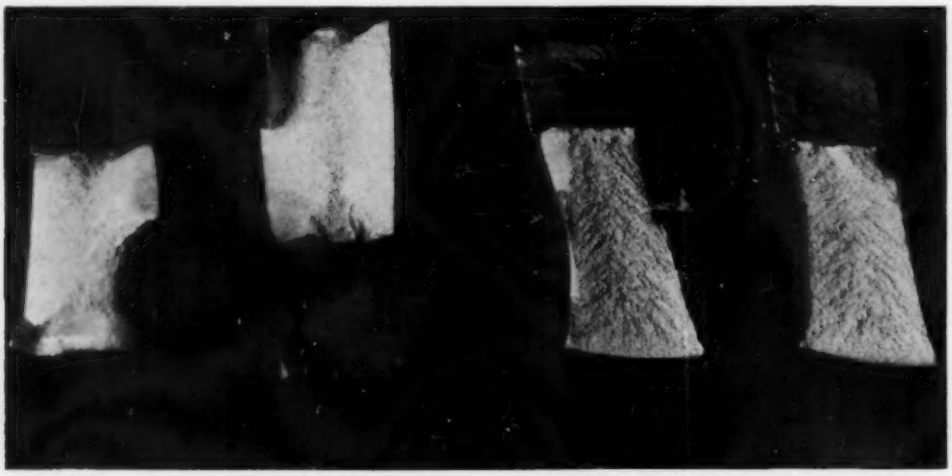
Material	*Spec. No.*	*Temp., of Test, °C.*	*Section at Notch, in.*	*Maximum Load, tons*	*Maximum Stress, tons/sq.in.*	*Elongation on 2 in., %*	*Reduction in Thickness, in.*	*Fracture Appearance*
Alloy *A*	T7	0	1·250×0·735	12·7	13·8	12·5	0·610	Single oblique shear with small mid-zone normal to plate surface (Fig. 22*a*)
	T8	−60	1·250×0·735	14·3	15·6	12·5	0·615	As T1 (Fig. 22*b*)
	T9	−100	1·250×0·735	14·8	16·1	14	0·595	As T1—slight increase of normal mid-zone (Fig. 22*c*)
	T10	−150	1·250×0·735	16·25	17·7	11·5	0·645	Mainly normal to plate surface (Fig. 22*d*)
	T11	−196	1·250×0·735	15·9	17·3	8	0·635	Almost completely normal to plate surface (Fig. 22*e*)
	T12	−196*	1·250×0·535	13·45	20·1	5	0·500	Completely normal to plate surface with some fissures (Fig. 24)
Alloy *B*	T1	0	1·255×0·750	16·7	17·75	12	0·690	Single oblique shear (Fig. 23*a*)
	T2	−60	1·255×0·750	16·9	17·95	11	0·650	Single oblique shear with small mid-zone normal to plate surface (Fig. 23*b*)
	T9	−100	1·255×0·750	17·9	19·0	13	0·650	Double oblique shear with larger mid-zone normal to plate surface (Fig. 23*c*)
	T4	−125	1·255×0·750	18·55	19·7	11·5	0·655	As T9 with fine fissures (Fig. 23*d*)
	T5	−150	1·255×0·750	18·35	19·5	9	0·650	Normal to plate surface with fissures (Fig. 23*e*)
	T6	−196	1·255×0·750	18·9	20·1	9·5	0·680	Normal to plate surface with pronounced fissures (Fig. 23*f*)
NE.5.M	T15	+22	1·240×0·760	14·35	15·2	13	0·625	Single oblique shear (Fig. 25*a*)
	T16	−150	1·250×0·700	16·75	17·4	19	0·590	As T15
	T17	−196	1·250×0·765	18·0	18·8	13	0·610	Normal to plate surface with some fissures (Fig. 25*b*)
5083	T13	+23	1·240×0·750	17·3	18·6	14	0·640	Single oblique shear (Fig. 26*a*)
	T14	−196	1·240×0·750	19·95	21·5	6	0·675	Completely normal to plate surface with pronounced fissure (Fig. 26*b*)
HE.9.WP	T18	+22	1·250×0·765	15·4	16·1	10	0·650	Single oblique shear (Fig. 27*a*)
	T19	−196	1·250×0·760	20·15	21·1	8	0·670	Double oblique shear (Fig. 27*b*)

* Notched on all four surfaces of the test specimen

with decrease of temperature. For alloys *A* and *C*, and the $\frac{5}{16}$ in. thick 5356 alloy, the load to initiate fracture increased with decrease of temperature down to −120° to −150°C. and appeared to remain constant with further decrease of temperature to −196°C. For alloys *B*, 5083, and extruded H9, all $\frac{3}{4}$ in. thick, the load to fracture increased to a maximum within the temperature range −160° to −196°C.

(2) The total energy values to fracture the specimens were greatest in the intermediate temperature ranges −120° to −150°C. compared with those at room temperature or −196°C. This tendency was repeated in the separate energies to initiate or propagate fracture.

For the 5083 alloy, the total energy was greatest at room temperature. This is considered to be associated



(a) (b) (c) (d)

12—Fracture appearance of Navy tear tests on hot rolled alloy *A* tested at: (*a*) +16°C., (*b*) −120°C., (*c*) −160°C., (*d*) −196°C.

(a) (b)

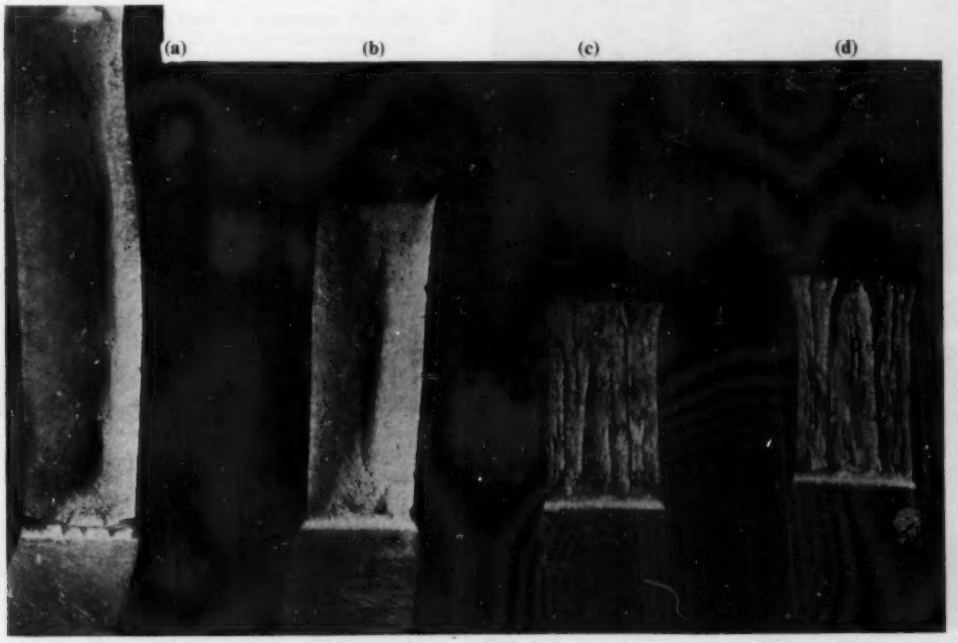
13—Fracture appearance of Navy tear tests on annealed alloy *A* tested at: (*a*) +16°C., (*b*) −196°C.

(a) (b)

15—Fracture appearance of Navy tear tests on annealed alloy *B* tested at: (*a*) +16°C., (*b*) −196°C.

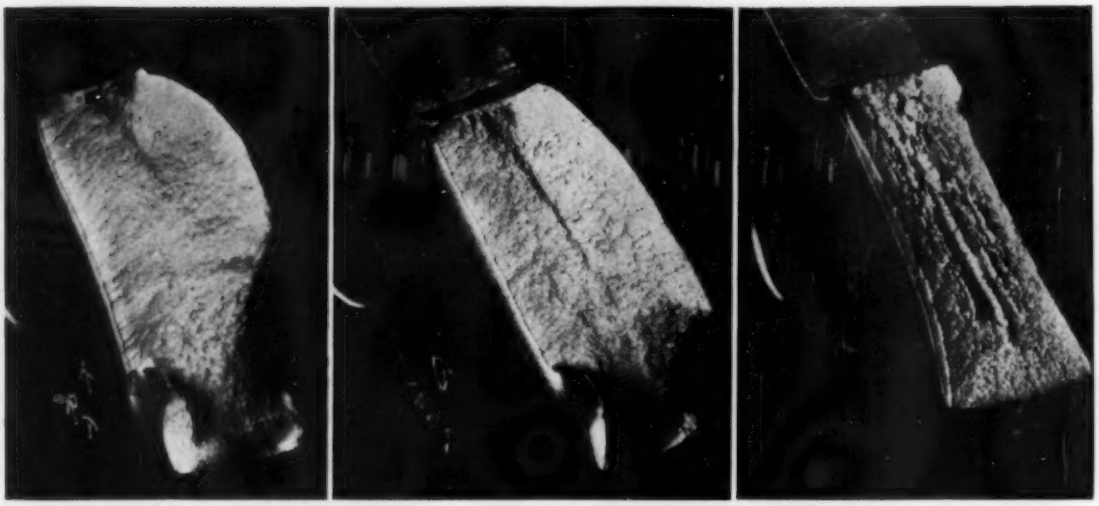
(a) (b) (c)

16—Fracture appearance of Navy tear tests on hot rolled alloy *C* tested at: (*a*) +16°C., (*b*) −120°C., (*c*) −196°C.



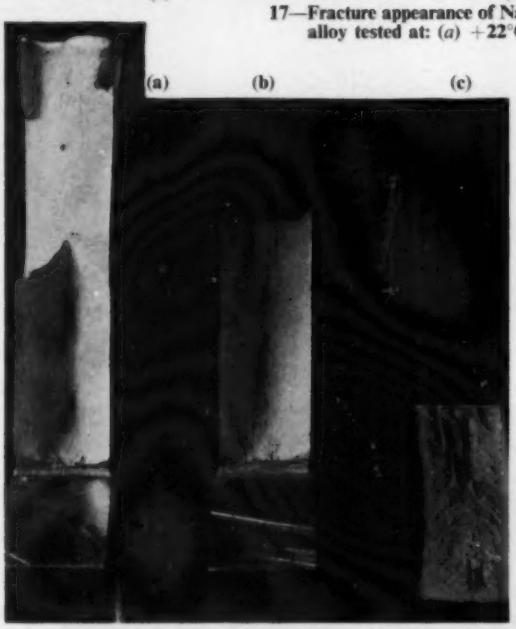
(a) (b) (c) (d)

14—Fracture appearance of Navy tear tests on hot rolled alloy *B* tested at: (*a*) +16°C., (*b*) −120°C., (*c*) −160°C., (*d*) −196°C.



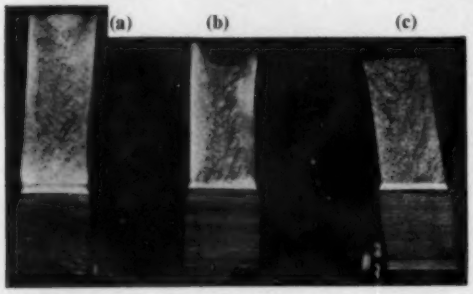
(a) (b) (c)

17—Fracture appearance of Navy tear tests on extruded NE.5.M alloy tested at: (*a*) +22°C., (*b*) - 150°C., (*c*) −196°C.



(a) (b) (c)

18—Fracture appearance of Navy tear tests on hot rolled 5083 alloy tested at: (*a*) +16°C., (*b*) −120°C., (*c*) −196°C.

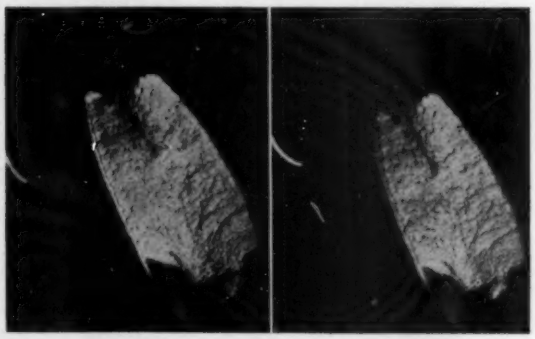


(a) (b) (c)

19—Fracture appearance of Navy tear tests on annealed 5083 alloy tested at: (*a*) +16°C., (*b*) −150°C., (*c*) −196°C.

(a) (b) (c) (d)

20—Fracture appearance of Navy tear tests on hot rolled 5356 alloy tested at: (*a*) +16°C., (*b*) −120°C., (*c*) −150°C., (*d*) −196°C.



(a) (b)

21—Fracture appearance of Navy tear tests on extruded NE.5.M alloy tested at: (*a*) +22°C., (*b*) −196°C.

with propagation of the fracture at an angle of approximately 45° to the longitudinal edge of the specimen, in tests at room temperature only (Fig. 18). This mode of fracture also occurred in tests on certain other alloys at room temperature (Figs. 13 and 14).

(3) The proportion of the total energies required to propagate fracture in the $\frac{3}{4}$ in. thick specimens,

generally showed minimum values at the intermediate temperature range −150° to −160°C. (Table II). For the $\frac{5}{16}$ in. thick 5356 alloy, the crack propagation energy increased with decrease of temperature to −196°C.

(4) The total energies and the energies to propagate fracture in the extruded NE.5 alloy were considerably greater than those for the hot rolled material, although the values of the maximum load to fracture were of the same order.

(5) The ductility of the hot rolled materials (Table II), were generally higher at a temperature of about −150°C. than at either room temperature or at −196°C. The ductility of the extruded NE.5 alloy increased with decrease of testing temperature to −196°C. The limited results for the extruded HE.9 alloy showed a higher ductility at room temperature than at −196°C.

The annealing treatment on specimens of hot rolled materials produced no general increase of ductility compared with the hot rolled condition.

(6) With the exception of the extruded HE.9 alloy, the alloys showed changes in the appearance of the fracture surfaces as the temperature of testing decreased to −196°C. The shear fracture oblique to the plate surface, and typical of rupture in tests at room temperature, progressively changed to a fracture completely normal to the plate surface at the lower

Table IV

Comparison of maximum load and reduction in thickness values for Navy tear test and Tipper tests

	NAVY TEAR TEST			*TIPPER TEST*		
Temperature of Test, °C.	*Spec. No.*	*Ratio:* $\frac{\text{Max. Load at T°C.}}{\text{Max. Load at R.T.}}$ %	*Reduction in Plate Thickness,* %	*Spec. No.*	*Ratio:* $\frac{\text{Max. Load at T°C.}}{\text{Max. Load at R.T.}}$ %	*Reduction in Plate Thickness* %
	Alloy A—Hot Rolled					
R.T.	A1	100	13	—	—	—
0	—	—	—	T7	100	17
−60	—	—	—	T8	112·5	16·5
−100	—	—	—	T9	116·5	19
−120	A2	116	10	—	—	—
−150	—	—	—	T10	128	12
−160	A3	122	20	—	—	—
−196	A4	122·5	15·5	T11	125	13·5
	Alloy A—Annealed					
R.T.	A5	100	17	—	—	—
−196	A6	108	20	—	—	—
	Alloy B—Hot Rolled					
R.T.	B1	100	10·5	—	—	—
0	—	—	—	T1	100	9
−60	—	—	—	T2	101	13
−100	—	—	—	T3	107	13
−120	B2	112	14·5	—	—	—
−125	—	—	—	T4	111	13
−150	—	—	—	T5	110	13
−160	B3	120	10·5	—	—	—
−196	B4	118·5	10·5	T6	113	9·5
	Alloy B—Annealed					
R.T.	B6	100	9·5	—	—	—
−196	B7	122	15	—	—	—
	Alloy C—Hot Rolled					
R.T.	C1	100	6·5	—	—	—
−120	C2	118	12·5	—	—	—
−196	C3	119	10·5	—	—	—
	NE.5.M—Extruded					
R.T.	F1	100	13·3	T15	100	17·7
−150	F2	118	17·1	T16	117	23·4
−196	F3	128	22·4	T17	125	20·2
	5083—Hot Rolled					
R.T.	D1	100	10·5	T13	100	14·5
−120	D2	107	12·5	—	—	—
−196	D3	114	15·5	T14	115	12·5
	5083—Annealed					
R.T.	D4	100	10	—	—	—
−150	D5	121	14	—	—	—
−196	D6	116·5	10·5	—	—	—
	5356—Hot Rolled					
R.T.	E1	100	8·5	—	—	—
−120	E2	108	12·5	—	—	—
−150	E3	108	16·5	—	—	—
−196	E4	109	14·5	—	—	—
	HE.9.WP—Extruded, Solution and Precipitation Treated					
R.T.	G1	100	14·5	T18	100	15
−196	G2	131	11·8	T19	131	11·8

(a) (b) (c) (d) (e)

22—Fracture appearance of Tipper notched tensile tests on hot rolled alloy *A* tested at: (*a*) 0°C., (*b*) −60°C., (*c*) −100°C., (*d*) −150°C., (*e*) −196°C.

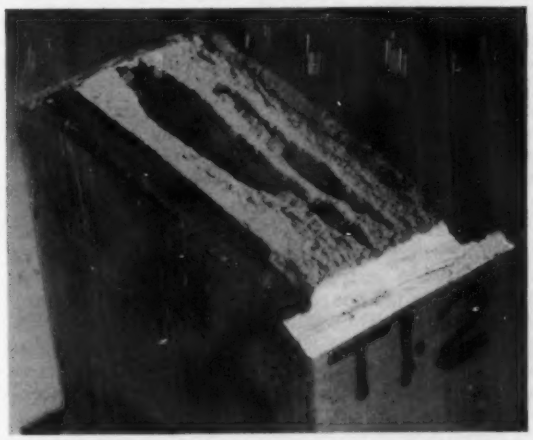
(a) (b) (c) (d)

(e) (f)

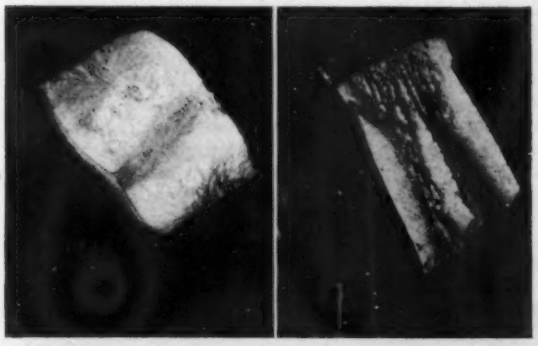
23—Fracture appearance of Tipper notched tensile tests on hot rolled alloy *B* tested at: (*a*) −100°C., (*b*) −125°C., (*c*) −150°C., (*d*) −196°C., (*e*) 0°C., (*f*) −60°C.

temperatures. The extruded HE.9 alloy showed single oblique shear characteristics throughout the whole range of testing temperatures.

In the hot rolled alloy *A*, the single oblique fracture changed to completely normal below −160°C., and a

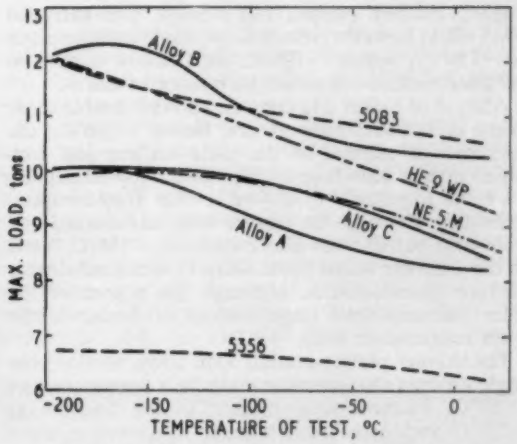


24—Fracture appearance of Tipper notched tensile tests on hot rolled alloy *A* tested at −196°C.

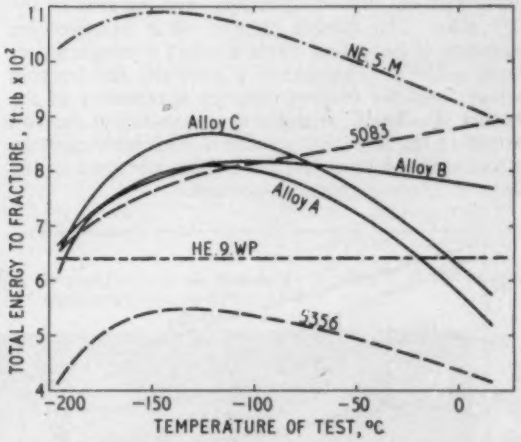


(a) (b)

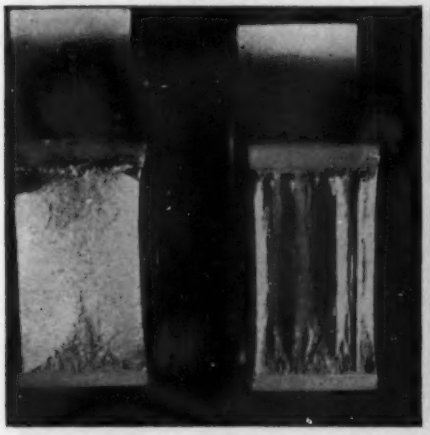
25—Fracture appearance of Tipper notched tensile tests on extruded NE.5.M alloy tested at: (*a*) +22°C., (*b*) −196°C.



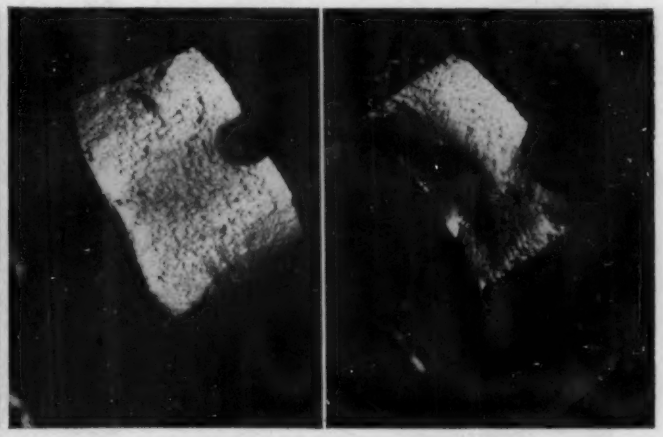
28—Relationship between maximum load and temperature in Navy tear tests



29—Relationship between total energy to fracture and temperature in Navy tear tests



26—Fracture appearance of Tipper notched tensile tests on hot rolled 5083 alloy tested at: (*a*) +23°C., (*b*) −196°C.



(a) (b)

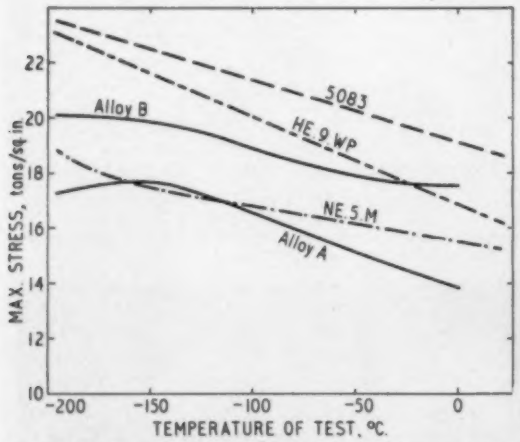
27—Fracture appearance of Tipper notched tensile tests on extruded HE.9.WP alloy tested at: (*a*) +22°C., (*b*) −196°C.

marked chevron pattern was evident. The extruded NE.5 alloy, however, retained the single oblique plane at −150°C., but at −196°C. the fracture normal to the plate surface contained pronounced fissures.

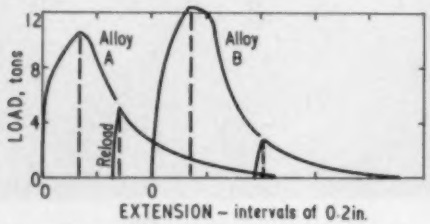
Alloy *B* of higher Mg content, showed double shear fractures at room temperature. Below −150°C., the fracture was normal to the plate surface and contained pronounced fissures. The 5083 alloy containing a similar Mg content, showed similar fracture characteristics, although the fissures were less marked and a chevron pattern was also evident at −196°C. Tests on the 1 in. hot rolled plate, alloy *C*, indicated similar fracture characteristics, although the planes of the shear fractures were single instead of double in the room temperature tests.

The thinner plate material, 5356 alloy, retained the single oblique characteristic down to a temperature of −150°C. Fissures were evident in the fracture at −120°C. and increased in intensity as the temperature decreased to −196°C.

(7) Annealing treatment carried out on three of the hot rolled alloys produced no regular variation in the Navy tear test results. The only change of fracture characteristics, due to annealing, was evident in the 5083 alloy. The double oblique shear obtained on specimens of hot rolled material tested at temperatures above −150°C., changed to a generally flat fracture surface; and the fissured chevron appearance of the fracture at −196°C. changed to a pronounced chevron pattern in the annealed specimen. This later chevron pattern differed however, from those obtained in the alloys of lower magnesium content.



30—Relationship between maximum stress and temperature in Tipper notched tensile tests



31—Navy tear tests load/extension curves for interrupted straining during fracture. Tested at −196°C.

Tipper notched tensile tests

(1) The general increases in load to fracture of the hot rolled and extruded alloys with decrease of temperature of the test, summarized graphically in Fig. 30, are similar to those obtained in the Navy tear tests. The maximum value of load to fracture for the hot rolled alloy *A* was obtained at the intermediate temperature of −150°C. For the higher magnesium alloys *B* and 5083, the fracture loads were greater, and the maximum values were recorded at −196°C. The extruded NE.5 alloy showed a similar increase of load to fracture to that of the hot rolled alloy *A*, although a maximum value was obtained at −196°C. compared with an intermediate temperature for alloy *A*.

(2) There were no severe reductions in the ductility values of any of the alloys tested at low temperatures. Hot rolled alloy *A* showed maximum ductility at −100°C. The extruded NE.5 alloy had greater ductility than the hot rolled material with a maximum indicated at −150°C. The higher magnesium alloys *B* and 5083, and extruded HE.9 showed some decrease of ductility at −196°C., which was probably associated with the presence of fissures on the fracture surfaces.

(3) The changes of fracture with decrease of temperature of testing are in agreement with those observed in the Navy tear tests. The fracture of the hot rolled alloy *A* at 0°C. shows single oblique shear, and with decrease of testing temperature a mid-zone normal to the plate surface increased in width to almost the full plate thickness at −196°C. Increasing the notch severity, by machining notches on all four surfaces of the specimen, increased the breaking stress, reduced ductility, and produced some fissures on the fracture surface. The extruded NE.5 alloy showed a change from the single oblique shear at room temperature to a fracture completely normal to the plate surface and containing fissures at −196°C.

The higher magnesium alloys *B* and 5083, had a single oblique shear fracture at room temperature,

32—Fractures in (*a*) alloy *A*; (*b*) alloy *B*, showing mode of propagation

which changed to double oblique at −100° to −125°C. followed by a fracture normal to the plate surface and with increasing severity of fissures as the temperature decreased to −196°C.

The extruded HE.9 alloy showed a change from single shear at room temperature to double oblique shear at −196°C.

STUDY OF THE CHARACTER OF FRACTURES

The fractures in the tear tests can be roughly classified as follows:

(i) Shearing type, in which the fractured surfaces are at about 45° to the plane of the plate surfaces. The fracture surface may consist of a single 45° plane, as in Figs. 12*a* and *b* or of two intersecting 45° planes, forming a vee, as in Figs. 14*a* and *b*.

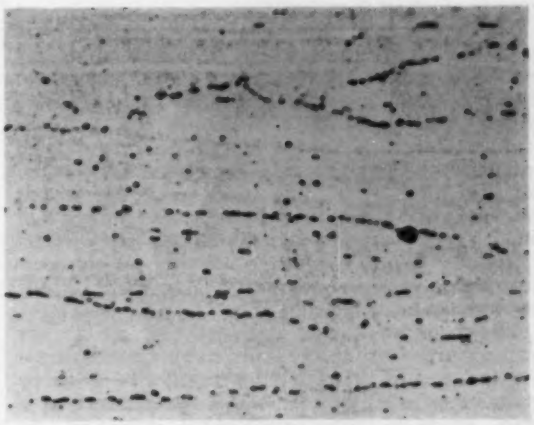
(ii) A generally flat fracture surface, perpendicular to the plate surface with borders of 45° planes, as in Figs. 16*a* and *b* and Figs. 19*a* and *b*. The border planes may be symmetrical, as in Fig. 19*a* or may run in opposite directions, as in Fig. 16*b*.

(iii) A generally flat fracture surface, perpendicular to the plate surfaces, with little or no evidence of borders, as in Figs. 12*d*, 13*b*, and 19*c*. This type is characterised by a chevron pattern of somewhat similar character to that found in brittle fractures of mild steel, but with an important difference which will be referred to later. Some of these fractures show longitudinal fissures, which can be clearly seen in Figs. 14*c* and *d*, 17*c*, and 20*d* and to which further reference will be made.

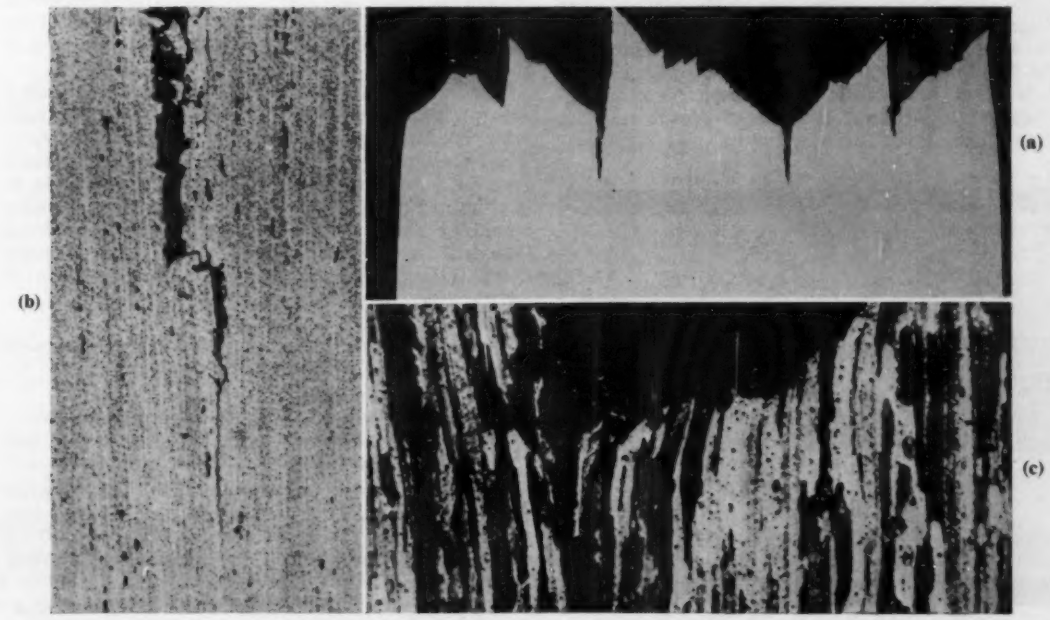
The foregoing types of fracture are in roughly decreasing order of energy absorbed in fracturing, as can be seen from a study of the load/extension diagrams, and the areas as given in Table II. From this it can be observed that the greatest changes in energy occur in the part related to propagation; *i.e.*, after maximum load.

All the fractures in the tear test specimens propagated slowly, the rate of extension during fracturing being of the order of $\frac{1}{10}$ in./min. It was found that the fracture could be arrested at any stage of the test merely by releasing the load. This behaviour is radically different from that observed in similar tests on mild steel plate, in which the fractures, if of the 'flat' type, progress very rapidly.

Reverting to the chevron pattern already noted, an investigation was made to ascertain the shape of the fracture front. At some interval during the fracturing of specimens of alloy *A* and *B*, the load on the specimen was quickly removed. The specimens were taken from the testing apparatus, and a red dye was poured into the cracks. The specimens were then quickly dried and returned to the apparatus, where the tests were completed. The load/extension diagrams for these interrupted tests are given in Fig. 31. It was found



34—Beta-phase at grain boundaries of alloy *B* etched in 10% aqueous phosphoric acid ×1350



33—Navy tear test on alloy *B* at −196°C.: (*a*) unetched section ×5; (*b*) longitudinal tear, etched in ½% hydrofluoric acid ×150; (*c*) longitudinal tear, electrolytically etched in phosphoric acid ×150

that the fracture fronts, as shown by the dye, were vee-shaped and pointed in the direction of propagation. For specimen alloy *A*, the vee embraced the whole plate thickness, whereas in alloy *B* the front consisted of a series of small vees, each bounded by a fissure (*see* Fig. 32).

On re-examination, the 'chevrons' were found to be roughly straight lines, perpendicular to the limbs of the vee fracture fronts. This configuration is different from that observed in brittle fractures of mild steel,[5] in which the chevrons are curved and the fracture fronts are of roughly parabolic form.

None of the fractures showed any evidence of 'crystallinity' or cleavage, all being of a mat texture indicating considerable deformation of the crystals. There was considerable reduction of plate thickness, even in the 'flat' fractures.

The longitudinal fissures referred to earlier were examined and were found to be unrelated to any macroscopical characteristic of the alloy structure, but evidently open up under transverse stresses arising between and at the apices of the separate small vee fracture fronts. A section across the fracture and specimen thickness, showed that the metal between the tears fractured by shearing at 45° to the tensile axis (Fig. 33*a*). The structure of the metal also showed that the fissures extend as cracks which were inter-crystalline (Fig. 33*b*) and may continue for consider-able distances within darker etching zones of the macrostructure of the alloy (Fig. 33*c*).

The mode of fracture suggested that, at the low temperature of testing, the transverse properties across the thickness of the material were affected either by grain boundary weakness or by the presence of a grain boundary constituent which influenced the nature of fracture with the increased yield strength at the low temperatures. A discontinuous boundary precipitation in the alloy has been identified as β-phase, Al_3Mg_2 (Fig. 34), which frequently occurs from ageing phenomenon in the alloys containing about 5% Mg.

A member firm of the Aluminium Development Association carried out a series of tensile tests on specimens cut from the longitudinal and transverse directions, and through the thickness of the plate. In spite of the occurrence of longitudinal fissures in the fractures at low temperature, the ductility values were the same or even greater than at room temperature. The results were as follows:

	Longitudinal		*Transverse*	
Temperature of Test, °C.	*Maximum Stress, tons/sq.in.*	*Elong., %*	*Maximum Stress, tons/sq.in.*	*Elong., %*
	(Averages of three tests at each temperature)			
+20	22·2	47	20·4	30
−70	22·3	53·5	21·0	32
−196	30·4	59	26·9	27
−196*	29·3	68	26·4	31

* Annealed at 400°C. for four hours, water quenched

Temperature of Test, °C.	*Tensometer Tests in 'Short Transverse' Direction*		
	Proof Stress	*Maximum Stress*	*Elongation, %*
	tons/sq.in.		
+16	11·5	19·4	10
−180	11·9	19·0	15
−180	12·7	23·4	20
−180	13·5	24·6	16
−180	11·9	25·4	20

TESTS ON WELDED SPECIMENS

The investigation was extended to determine the properties of metal-arc butt-weld deposits on the $\frac{3}{4}$ in. thick NP.5/6 alloy plates. Test pieces 15–20 in. long were prepared for welding. The edges of the plates were machined in a symmetrical double-**U** contour with a $\frac{1}{8}$ in. root face. The plates were pre-heated to 200°C. before welding.

For the initial deposit, the plates were set-up vertically. The root run was made by the simultaneous double-operative argonarc process, using $\frac{3}{16}$ in. filler wire with 5% Mg (NG.6 quality). The filler wire was fed from one side of the plate only. This technique[6] avoids the need for back chipping and maintains a large weld pool across the root of the butts, thus giving less chance for any oxide slag to be trapped at this position.

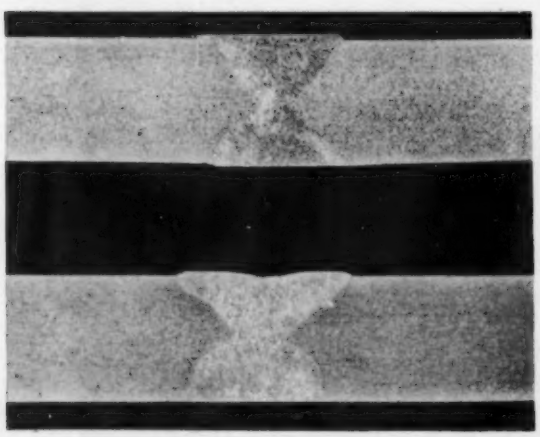
The filling runs of weld metal were deposited separately on each side of the plate, using the self-adjusting argonarc process in the downhand position. The filler wire was 16 s.w.g. of NG.6 quality, operated on a d.c. supply of 200 amp with the plate negative. It is estimated that the loss of magnesium from the wire would be about 0·5–0·75%. Such welds have been repeatedly shown to be of 95% efficiency under tensile testing. The appearances of the macrostructures on transverse sections through the weld deposits are shown in Fig. 35.

It was decided to investigate the characteristics of three qualities of weld deposit:

(i) Sound welds free from oxide films
(ii) Welds containing a controlled porosity of about 10%
(iii) Inferior weld deposits containing some porosity and oxide films, and showing some lack of penetration.

The weld deposits were subjected to radiographic and ultrasonic examinations. It was noted that radiography, in some instances with a sensitivity of 0·7%, did not reveal some defects which appeared on the fracture surfaces.

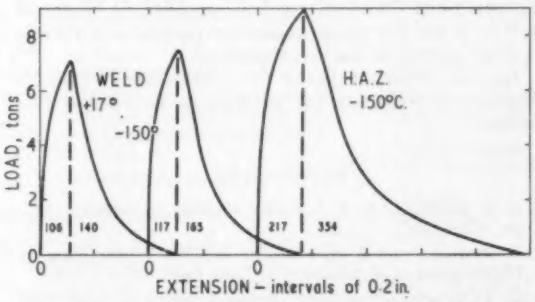
Each Navy tear test specimen was notched in either the weld metal or the heat-affected zone, and the deposit was normal to the direction of the tensile load.



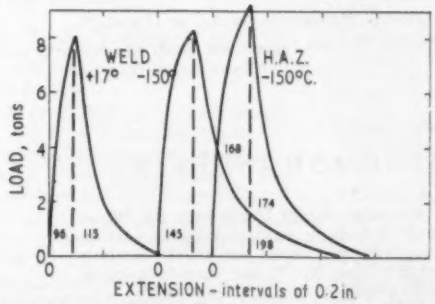
35—Macrostructure of sound welds: Alloy *A* (top); alloy *B* (bottom)

Table V
Navy tear test on welded specimens

Material	Weld Quality	Notch Position	Temp. of Test, °C.	Max. Load, tons	Corrected Energy Values, ft.lb Initiation	Propagation	Total	Ratio: Energy to propagate / Total Energy %	Reduced Plate Thickness, in.	Fracture Appearance
Alloy *A*										
WA	Sound	In weld	+17	7·1	106	140	246	56·9	0·675	Slight chevron pattern and 'cokey' appearance (Fig. 39*a*)
WB	,,	,,	−150	7·5	117	165	282	58·5	0·675	As WA
WC	,,	H.A.Z.	−150	8·95	217	354	671	52·7	0·660	Partially shear and 'cokey' (Fig. 39*c*)
WD	Oxide films	In weld	−150	6·7	84	138	222	62·2	0·650	'Cokey' with gross discontinuity (Fig. 39*b*)
Alloy *B*										
WE	Sound	,,	+17	8·05	96	115	211	54·5	0·685	As WA (Fig. 40*a*)
WF	,,	,,	−150	8·25	145	198	343	57·8	0·705	As WA
WG	,,	H.A.Z.	−150	9·2	168	174	342	50·8	0·680	As WC (Fig. 40*b*)
WH	Porous	In weld	−150	8·4	146	190	336	56·6	0·670	'Cokey' and porous (Fig. 40*c*)
WJ	Oxide films	,,	−150	9·1	149	181	330	63·5	0·660	'Cokey' with discontinuities (Fig. 40*d*)



36—Navy tear test load/extension curves for sound weld deposits in NP.5/6 alloy *A*, using NG.6 filler wire



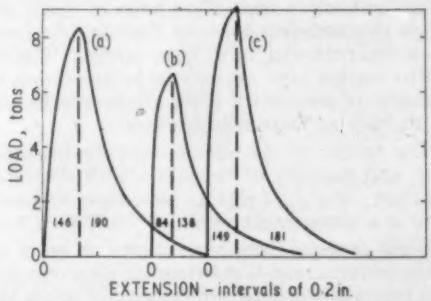
37—Navy tear test load/extension curves for sound weld deposits in NP.5/6 alloy *B*, using NG.6 filler wire

The tests were carried out at room temperature and at −150°C. The results of these tests and the energy values obtained from the load/extension diagrams (Figs. 36–38) are contained in Table V.

DISCUSSION OF RESULTS ON WELDED SPECIMENS

(*a*) In spite of the variation of weld metal quality, the values for load to fracture and the reduction of plate thickness at fracture were generally similar.

(*b*) The values for maximum load were lower for the weld deposits than for the parent materials. At



38—Navy tear test load/extension curves for defective weld deposits in NP.5/6 alloys using NG.6 filler wire tested at −150°C.: (*a*) Alloy *B*, weld metal porous; (*b*) alloy *A*, oxide film in weld deposit; (*c*) alloy *B*, oxide films in weld deposit

room temperature, the decrease in load to fracture for the welds in alloys *A* and *B*, were 13·5 and 21·5% respectively, whilst at a temperature of −150°C. these values increased to 25 and 33% respectively.

(*c*) The welds in the higher magnesium alloy *B* were of greater strength than those in alloy *A*.

(*d*) The increase of maximum load at the lower temperature, compared with that at room temperature, was less than that obtained on the unwelded specimens.

(*e*) The energy values for initiation and propagation of fracture are considerably lower than those obtained on unwelded material.

(*f*) A comparison of the results of tests on welds of qualities *a* and *b* showed no appreciable differences in the energy values.

(*g*) Extensive defects in specimen WD were responsible for low values of maximum load and initiation energy.

(*h*) All the fractures have a dull 'cokey' appearance and with evidence of a slight chevron pattern.

CONCLUSIONS

(1) Static tensile tests on notched specimens of aluminium–magnesium alloys to BS. 1477 NP.5/6,

(a) (b) (c)

39—Fracture appearance of Navy tear tests on welded specimens of alloy *A* using NG.6 filler wire: (*a*) WA; (*b*) WD; (*c*) WC

(a) (b) (c) (d)

40—Fracture appearance of Navy tear tests on welded specimens of alloy *B* using NG.6 filler wire: (*a*) WE; (*b*) WG; (*c*) WH; (*d*) WJ

BS. 1476 NE.5 and HE.9 and to American Designation 5083 and 5356, have shown changes in the appearance of the fractures with decrease of testing temperature from ambient to the boiling point of liquid nitrogen. Two characteristic types of fracture obtained at very low temperatures have been studied. The tests show that neither type appears to be associated with or indicative of a tendency of the alloys to behave in a brittle manner at these temperatures.

(2) The results of the test confirm the increase of strength and ductility of the alloys with decrease of temperature, and in general, maximum values are attained at a temperature of about −150°C.

(3) Weld deposits have been shown to be of considerably lower strength than parent plate materials, and the increase of strength of the deposit at the lower temperature of testing is only small.

ACKNOWLEDGMENTS

The author wishes to thank the Aluminium Development Association and member firms, particularly Messrs. James Booth & Co. Ltd. who carried out the tensile tests reported in the section on the study of the fractures; also the Directors of British Oxygen Research and Development Ltd., and Mr. D. Slater of A.P.V. Ltd., for their assistance and advice during various phases of the investigation.

He also wishes to thank the Committee of Lloyd's Register of Shipping for permission to publish this paper.

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TECHNICAL LIAISON — RESEARCH AND INDUSTRY

For the small and medium-size engineering and industrial firms which are unable to support a research team or to employ a number of qualified technologists, it is often a serious difficulty to obtain answers to specific problems in manufacture. One of the main purposes of the research organizations set up under the Department of Scientific and Industrial Research is to provide such firms with facilities for co-operative research and a means of disseminating technical information. But how successful in fact are these research institutions?

To find out, the Scottish Council (Development and Industry) set up an experimental Technical Liaison Service with Conditional Aid Funds.

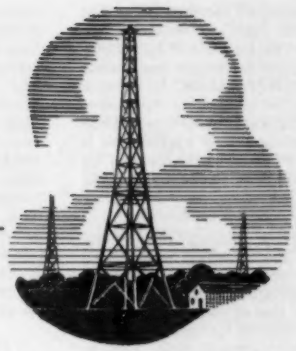
A team of three investigators – retired technical or managerial executives – visited over 100 industrial firms in Scotland to discuss their problems and, if possible, to find answers to them.

A report "Small Firms with Big Problems" has been published by the DSIR (Available free from the DSIR Lending Library Unit, 20 Chester Terrace, London W.1) giving the results of this survey.

It seems that technical publications are the main source of information on developments and that problems are answered mainly by suppliers and other firms in the same field. There is an obvious danger here that such a 'closed-circuit' of experience is likely to stultify development or applications in new fields.

Three research associations appeared to be outstandingly useful: BCIRA, BISRA, and BWRA, but in general it is evident that unless these organizations can set up local technical liaison services their written publicity and reports are quite inadequate for bridging the gap between them and the firms who need the information they can in fact provide.

NEWS
and Announcements



News of the Institute and Branches
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INSTITUTE ACTIVITIES

Autumn Meeting 1958

CORRECTION

The programme published in the October issue of the *Journal*, especially in regard to the order of papers to be presented at technical sessions, is incorrect.

Details as stated in the programme circulated to members are correct, except that the Works Visit on Thursday evening 6 November, has been cancelled.

The final programme is as follows:

Wednesday, 5th November

7.15 for 7.45 p.m.—Annual Dinner at the Park Lane Hotel.

Tickets inclusive of cocktails £3. Members are entitled to one ticket at £2 10*s*. on personal application.

Thursday, 6th November

10.0 a.m. to 12.30 p.m.—Presidential Address by Mr. J. Strong, M.A., followed by the presentation and discussion of the following papers:

"*Development of the Tungsten-Arc Cutting Process*" by R. A. Cresswell (Aug.)

"*Developments in the Inert-Gas Tungsten-Arc Fusion Spotwelding Process*" by F. W. Copleston and L. M. Gourd (Sept.)

"*Developments in Power Sources for Inert-Gas Tungsten-Arc Welding*" by M. Waller (Sept.)

"*New Developments in Arc Welding Processes and Their Applications*" by P. Shaw and D. B. Tait (Oct.)

2.0 to 4.0 p.m.—Presentation and discussion of the following papers:

"*Why Weld Automatically?*" by I. C. Fitch (July)

"*Flame Cutting and Jigging for a Nuclear Power Station*" by J. W. Gethin (Sept.)

"*Carbon Dioxide Welding and Production of Automobile Chassis Side Members*" by S. Astley (Sept.)

"*Positioners for Automatic Welding of Atomic Reactor Housings*" by W. A. Davis (Oct.)

4.0 p.m. Tea

Friday, 7th November

10.0 a.m. to 12.30 p.m.—Presentation and discussion of the following papers:

"*Metal Arc Welding of Aluminium Bronze Alloys*" by M. K. Williams and W. I. Pumphrey (Oct.)

"*Welding of Some Creep Resisting Steels*" by J. F. Mercer (Oct.)

"*Some Guidance on Welding Dissimilar Alloy Steels*" by M. C. T. Bystram (Oct.)

"*Manual 'Open Air' Welding of Reactive Metals*" by J. C. Borland and W. G. Hull (Sept.)

2.0 to 4.0 p.m.—Presentation and discussion of the following papers:

"*Current and Pressure Wave Shapes used in Resistance Welding*", by W. A. Knipe (Aug.)

"*Application of the Numerical Analogue System to Oyxgen Cutting Machines*" by L. J. Hancock and J. R. Arrowsmith (Aug.)

"*Some Aspects of Cross Wire Welding of Aluminium Alloys*" by G. H. Batten (Sept.)

"*Experiments with Self-Guiding Equipment for Aluminium*" by J. E. Tomlinson and R. V. Blewett (Oct.)

"*High-Current Inert-Gas Metal-Arc Welding of Aluminium*" by A. A. Smith and P. T. Houldcroft (Sept.)

All the papers for the technical sessions have been published in the July–October issues of the BRITISH WELDING JOURNAL, as indicated in parentheses.

The Principal Guests at the Dinner will be the Rt. Hon. Aubrey Jones, P.C., M.P., the Minister of Supply; the Rt. Hon. Lord Coleraine, P.C.; and the Rt. Revd. The Bishop of Kensington.

Hydrogen in Steel

A special joint meeting is being held with The Iron and Steel Institute, on Wednesday, **10th December, 1958,** to present and discuss the following papers:

Hydrogen Content of Mild and Alloy Steel Weld Deposits, by P. D. Blake (BWJ, March 1957, p. 146)

Measurement of the Diffusible, Residual, and Total Hydrogen Contents of Weld Metal, by P. D. Blake (BWJ, March 1958, p. 126)

Effect of Diffusible Hydrogen on the Mechanical Properties of Weld Metal, by P. D. Blake (BWJ, December 1958—to be published)

Hydrogen in Mild Steel Weld Deposits, by N. Christensen and R. Rose (BWJ, December 1955, p. 550)

Hydrogen in Mild Steel Weld Deposits, by N. Christensen, K. Gjermundsen, and R. Rose (BWJ, June 1958, p. 272)

The meeting will be held in the **Weir Lecture Hall** of the **Institution of Naval Architects,** 10 Upper Belgrave Street, London, S.W.1, and **not** at the Iron and Steel Institute as announced on the 'Meeting Card', and will commence at 2.30 p.m.

Welding in Low Temperatures

Dr. **D. L. Pratt,** Chief Engineer of the Trans-Antarctic Expedition, addressed a well attended meeting of the Institute at 54 Princes Gate on 24th September, on the subject '*Welding Repairs in Low Temperatures*'.

The chair was taken by Mr. John Strong, President of the Institute, and the proceedings opened with the showing of a film of the first voyage of the *Theron*. The subsequent lecture was illustrated by nearly 100 coloured transparencies, and covered all phases of the expedition, including the preparatory re-welding of the caterpillar tracks on the Sno-Cat tractors.

Mr. J. F. Lancaster proposed a vote of thanks and voiced the appreciation of all members present for a lecture of unusual interest.

School of Welding Technology

In view of the response to the special courses held between October 1957 and June 1958, the Institute is once again holding a series of 5-day welding courses as listed in the September issue of the *Journal*. From February 1959 it is planned to

increase the frequency of the courses and to hold twenty each year. Because of the heavy demand for places, it will be necessary, from January 1959, to give preference to members of the Institute and to the employees of Industrial Corporate Members. The number of places for each course is strictly limited, so that early application is essential. A copy of the prospectus may be obtained from the Secretary of the Institute.

C. W. Hill Prize

The Council invite competition for the 1958 C. W. Hill Prize, which is open to Graduates and Associates of the Institute.

Competitors are required to re-design for welded fabrication one of two designs to be supplied by the Institute, or any riveted or cast object of their own selection. Entries must consist of sufficiently detailed arrangement drawings, and must specify the materials candidates propose to employ in the design. Drawings must be accompanied by notes of up to 500 words describing and explaining the principles of approach for the design, with sketches or diagrams as considered necessary.

The designs to be supplied by the Institute are:

1. A bolted lattice girder.
2. A cast press bolster.

The closing date for the receipt of entries is 1st February 1959, and the prize, which will be awarded at the 1959 Spring Meeting, will consist of drawing instruments or books, or both, to the approximate value of £21.

Those wishing to compete should apply to the Secretary of the Institute for further particulars.

NEWS OF MEMBERS

Mr. **Morton Gordon** has been appointed acting chief welding engineer of Metal and Pipeline Endurance Ltd.

Mr. **S. H. Griffiths,** Head of the John Thompson Ltd. Group Research Laboratories, has been appointed a Director of John Thompson (Wolverhampton) Ltd. Mr. Griffiths is a member of Council of the Institute and is Past Chairman of the Wolverhampton Branch.

Mr. **T. Nellis,** formerly Welding Engineer with John Brown Land Boilers Ltd., has joined the staff of G. & J. Weir Ltd., Glasgow, as Welding Engineer, responsible for the Fabrication, Welding, Heat-Treatment and Smithy Departments.

Mr. **E. G. B. Trehearne** is now with the Development Department of Union Carbide Ltd.

The Council regret to record the deaths of the following members:

Mr. **F. C. Cocks,** B.Sc., previously a principal surveyor to Lloyds Register of Shipping. (Member 1943, North London Branch.)

Mr. **W. H. Porter** (Associate 1947, North London Branch).

BRANCH NEWS

East of Scotland

At the first meeting of the new session a visit was arranged to Euclid (Gt. Britain Ltd.), Newhouse, Lanarkshire.

A very cordial reception was given to all the members. A lecture and film described the types and applications of earth-moving equipment manufactured. They were followed by a tour of the works and the members were most impressed by the general layout of machines for shearing, punching and profiling of steel plates. This continuous layout continued to the fabrication shop where the majority of work is carried out. For steel fabrication both manual and automatic arc welding were extensively used. These layouts and the welding jigs employed had very favourable comments from the members. Tea was served following the Works' visit and question time was allowed. In all, 18 members visited the Works and wholeheartedly voiced their approval and their thanks. Another visit at an early date is probable. A.W.M.

CONTRIBUTORS TO THE JOURNAL

J. F. Mercer, A.I.M., Research Metallurgist at Murex Welding Processes Ltd., was born in 1923 at Llanelly, Carms. He was educated at Llanelly Grammar School and at Swansea Technical College. Mr. Mercer then received metallurgical training at Bynea Steel Works, Loughour, Swansea, and first entered the welding industry in 1946 as Works Metallurgist of United Metal Electrodes and Welding Equipment Ltd. In 1950 he joined Messrs. C.A.V., Diesel equipment manufacturers, as Senior Metallurgist. He joined his present company in 1952 and has been primarily concerned with the development of low-hydrogen electrodes and the welding of low-alloy and special purpose steels.

M. K. Williams, B.Sc., Research Investigator in charge of investigational and development work on non-ferrous welding at Murex Welding Processes Ltd. He was born in 1919 in Bridgend, Glamorganshire, and was educated at Bridgend County Grammar School and at University College, Cardiff. where, in 1941, he graduated with B.Sc. Metallurgy and Chemistry (Hons. Chem.). Mr. Williams has a wide experience of general metallurgical practice in light and precision engineering, melting, casting and manipulation of precious metals, powder metallurgy and joining of metals. He joined his present company in March 1957.

Mr. Williams and Mr. Mercer were authors of papers printed in the October issue of the *Journal*.

R. J. Durham, Investigator in the Engineering Division of Aluminium Laboratories, was educated at Farnham Grammar School and later at Oxford College of Technology. He joined his present Company in 1946, and is now engaged on research into the mechanical properties of aluminium alloys.

BRITISH WELDING RESEARCH ASSOCIATION

Annual General Meeting—1958

Three new Council members were elected at the Annual General Meeting of the Research Association, held on 25th September. They are Dr. R. Beeching (Technical Director on the main board of I.C.I. Ltd.), Mr. J. M. Willey (Director and General Manager of Murex Welding Processes Ltd.), Mr. T. M. Herbert (Director of Research, British Railways Division of the British Transport Commission). Mr. J. Strong, Executive Director of British Oxygen Gases Ltd., and this year's President of the Institute of Welding is the Institute's representative on Council.

Staff changes—London Office

Mr. Bernard E. Dorey, A.M.I.E.E. has recently joined the staff as a Development Officer. He was previously Technical Manager of the Plant Laboratory at Arc Manufacturing Co. Ltd. From 1937 to 1945 he was with Quasi-Arc Limited and the British Oxygen Co. Ltd. He has been closely associated with the development of submerged-arc, argon-arc, and self-adjusting inert-gas welding equipment in Britain.

Mrs. Joan Lindley, B.Sc., Publications Officer, who joined the staff in 1956, has left to accompany her husband on a temporary appointment in the North of England. Her work has been taken over by Mr. John Scott, B.Sc., A.M.I.Struct.E., who recently left the staff of *Iron and Coal Trades Review* to become Press Officer of the Association.

Miss Audrey D. Smith, B.A., has recently joined the Publications Department. Her first language is French, and this will be of considerable value since much of the Department's work deals with foreign publications.

NEWS FROM INDUSTRY

Courses on metalwork fabrication

Production Engineering Ltd. will hold, at the Glasgow headquarters of the Institution of Engineers & Shipbuilders of Scotland, a series of one day courses for foremen and junior executives of Scottish engineering firms. The courses are to be held monthly from September until June next and will include among the subjects to be studied the latest developments in welded steelwork fabrication.

J. F. Mercer

M. K. Williams

R. J. Durham

BEAMA Conference

The British Electrical & Allied Manufacturers Association (BEAMA) are organizing a Publicity Conference at the Connaught Rooms, London, W.C.2 on 11th and 12th November. Attendance is open to BEAMA member firms only. Full details may be obtained from the organizers at 36 and 38 Kingsway, London, W.C.2.

Powder fluxes for welding

Under the name of Muraflux, Murex Welding Processes Ltd. are producing a group of new powder fluxes for submerged-arc welding. The first of these, Muraflux A, is a high quality general purpose flux for submerged-arc welding of mild steel and can be used on either a.c. or d.c. supplies with welding currents up to 900 amp.

In conjunction with Murawire the flux has been approved by Lloyds Register of Shipping and is accepted by the Ministry of Transport.

Mild steel electrodes

An iron powder electrode known as Ferrolux, is being manufactured by Quasi-Arc Ltd. It has been designed for high speed welding of mild steel in the flat and horizontal positions, and is lightly tipped with a special compound to ensure instantaneous starting. The electrode has been approved by Lloyds Register of Shipping and by the Ministry of Transport & Civil Aviation.

The following results are typical of those obtained from test specimens welded with the electrode in accordance with BS. 639: 1952. yield, 29 tons/sq.in.; U.T.S., 34 tons/sq.in.; elongation on 3·54D, 28½%; reduction in area, 48%; Izod impact value (V-notch), 63 ft.lb; bend over 180° without failure.

Another addition to the rapidly increasing range of 'contact' type of electrodes has also been made by the Company with the introduction of the Mirrospeed electrode. In sizes ranging from 12–4 s.w.g. the electrodes have been designed for fast and economic welding of mild steel in all positions. They are particularly suitable for flat and horizontal-vertical fillet welding and flat butt welding. In fillet welding the electrode can be 'drawn out. to give very long run lengths. In addition, it is usual to use one size larger gauge than normal, so that even longer runs can be made at high speeds, and less time is lost in changing electrodes.

The electrodes conform to BS. 639:1952, and are approved by Lloyds Register of Shipping and the Ministry of Transport and Civil Aviation for welding mild steel in all positions.

Synchronous welding controllers

W. H. Sanders (Electronics) Ltd. are offering a range of synchronous controllers of which Type ICT09 is intended primarily for the control of seam and stitch welders and incorporates Ignitron Contactor Type IC10 and Weld Timer Type WT15. The contactor uses size 'C' ignitrons together with control thyratrons whose initial heating time is only 10 seconds.

The timer provides weld current durations of 2–30 cycles, adjustable in one cycle steps, repeated with an on-off time interval of 2–10 cycles. Heat variation over a range of approximately 50% can be obtained by the phase shift control provided.

There is a simple lead/trail action ensuring that weld current starts during the opposite half cycle from that in which the previous weld interval ended. No relays are used in the timing circuits, and particular attention has been paid to 'fail-safe' features.

A timer, Type WT16, is also available in which air valve control circuits are included for welding gun operation, with pressure switch or 'squeeze-time' current initiation.

The full range of equipment will include facilities for pulsation welding, current variation during weld (slope control), programmed timer operation for light alloy and heat resisting metals, and automatic current control; three-phase controllers using the same basic circuits are also being developed.

DIARY

3rd Nov.—West Wales—*Reclamation of Plant and Machinery* by G. C. Musted (Swansea Technical College).

4th Nov.—East Wales—*Reclamation of Heavy Plant and Machinery by Practical Welding Applications* by G. C. Musted (South Wales Institute of Engineering, Cardiff).

South Western—*The Welding of Tubular Structures* by G. B. Godfrey (Radiant House, Bristol 1, 7.15 p.m.)

5th Nov.—Manchester—*Residual Stresses; Do they matter?* by Dr. L. E. Benson (Manchester College of Science and Technology).

6th Nov.—North Eastern (Tyneside)—*Non-destructive Examination of Welded Structures* by T. V. Brown.

10th Nov.—Sheffield and District—*Repairs to Heavy C.I. Castings by the Gussolite Process* by D. H. Jones (Grand Hotel, Sheffield, 7.15 p.m.)

12th Nov.—Liverpool and District—*The Gas Cutting of Metal Plates* by Dr. L. C. Bannister. Combined meeting with the Liverpool Metallurgical Society (hosts). 7.15 p.m.

Wolverhampton—*Paper on Nuclear Welding in the Nuclear Field* by J. A. Forrest or J. McLean (Holly Bush Hotel, Wolverhampton, 7.30 p.m.).

13th Nov.—South London—*Welding and Brittle Fracture* by W. D. Biggs (54 Princes Gate, S.W.7., 7.30 p.m.)

14th Nov.—Eastern Counties—*Joining of Dissimilar Metals* by A. A. Smith and J. C. Young (Ipswich).

17th Nov.—Manchester—*Design for Welding* Q. and A. meeting. Panel—R. J. Fowler, E. Fuchs, Dr. F. Koenigsberger, W. L. Partridge (Manchester Engineers Club).

18th Nov.—East Midland—Members Papers (Loughborough College, 7.15 p.m.)

19th Nov.—South Western—*Practical Aspects of Resistance Welding* by Mr. McDowell and Mr. Weeks (Radiant House, Bristol 1, 7.15 p.m.)

West of Scotland—Visit to Colville's Works.

North London—*Fabrication of Sheet Metal using the Argon Arc Spot Welding Process* by F. Copleston (54 Princes Gate, S.W.7, 7.30 p.m.)

20th Nov.—Southern Counties—*The Welding of Pipes* (to be arranged by Stewarts & Lloyds Ltd.)

25th Nov.—West of Scotland—*Welding* by J. Watson. Joint Meeting with Institution of Structural Engineers.

26th Nov.—South London—*Some Aspects of Structural Welding* by E. M. Lewis (Brixton School of Building, 6.30 p.m.)

South London (Medway Section)—*Texas Towers* followed by a Discussion (Sun Hotel, Chatham, Kent, 7.30 p.m.)

East of Scotland—*Introductory Metallurgy* by Mr. L. M. Gourd, B.Sc.(Eng.) (Met.)

2nd Dec.—North London (Slough Section)—*Welding in Nuclear Engineering* by A. Prince (Community Centre, Slough).

3rd Dec.—Manchester—*Welding in the Rocket Motor Industry* by F. J. Wilkinson (Manchester College of Science and Technology).

North London—*Fatigue and Brittle Fracture Aspects of the Design of Plate Girders* by E. M. Lewis (54 Princes Gate, S.W.7, 7.30 p.m.)

4th Dec.—South Western—*Welding Forum.* Joint Meeting with Gloucester Assoc. of Engineering (Gloucester, 7.15 p.m.)

West Wales—*Automatic Welding* by Dr. W. E. Freeth (Neath Technical College).

North Eastern (Tyneside)—*The Use of Welding in Ship Repair Work* by A. Bowman.

Leeds and District—Annual Dinner (The Mansion Hotel, Roundhay, Leeds).

Eastern Counties—*The Plastic Theory of Structures* by T. M. Charlton (Norwich).

6th Dec.—North Eastern (Tyneside)—Branch Dinner.

STUB ENDS

▶ The Executive Offices of Kelvin & Hughes Ltd. have moved from 2 Caxton Street, London, S.W.1, to Empire Way, Wembley, Middlesex. Telephone: Wembley 8888.

▶ The American Welding Society has issued section II of the fourth edition of the Welding Handbook: Gas, Arc and Resistance Welding Processes.

▶ Sciaky Electric Welding Machines Ltd., Slough, have produced a new version of their S.F.15 foot operated spot welding machine, with a working throat of 30 in. and an electrode stroke adjustable to a maximum gap of 4 in.

▶ The Mine Safety Appliances Co. Ltd., Queenslie Industrial Estate, New Edinburgh Road, Glasgow, E.3, a subsidiary of the U.S.A. concern of the same name is increasing its range of safety products to include obnoxious gas alarms, and protective and safety equipment.

▶ A complete series of Oxweld CO_2 regulators has been introduced by the Linde Department, Union Carbide International Co., New York, U.S.A.

CLASSIFIED ADVERTISEMENTS

Two Welding Development Engineers required to join a specialized team engaged on development and research into new techniques of welding. Candidates for the first vacancy should be of degree standard with a sound metallurgical or engineering background, and with previous experience of welding development. H.N.C. in Metallurgy or Engineering is desirable for the second vacancy, together with familiarity of inert-gas arc processes.

The work will include welding of non-ferrous metals and high tensile steel, and the development of welding methods suitable for the manufacture of rocket motors and high pressure vessels. Applications in confidence are invited by: The Personnel Manager, Bristol Aircraft Limited, Weston Division, Oldmixon, Weston-super-Mare.

Current
WELDING LITERATURE

Book Reviews
Additions to the Institute Library
Trade Publications

CONTENTS OF PERIODICALS RECEIVED

THIS SECTION is intended to give a survey of the current welding literature received by the Institute of Welding Library. The contents lists are not exhaustive; only the main articles in welding journals are listed, and reprints from other journals and short notes are generally excluded. In addition, welding articles from other periodicals are listed. Abstracts of welding literature are given in the *Bibliographical Bulletin of Welding and Allied Processes*, published by the International Institute of Welding, and details of this may be obtained from the Secretary of the Institute of Welding.

Welding Journals

Schweissen und Schneiden (Germany), 1958, vol. 10, July

The metallurgical conditions for the welding of clad steel-plates, W. Rädeker (255–259)

Selected examples of welding structures in chemical engineering, R. Söhngen (260–265)

Application of welding in apparatus construction of the chemical industry, P. Richter (265–269)

Contribution to design and calculation of welded hollow bodies for high operation pressures and elevated temperatures, A. F. Maier (270–276)

Problems of control engineering for resistance welding machines W. Masing (276–281)

The application of wiping solder in the manufacture of motor car bodies, H. Johnen (281–283)

Welding Journal (U.S.A.), 1958, vol. 37, July

Inert-gas tungsten-arc welding of titanium for nuclear and chemical industries, G. M. Adamson and W. J. Leonard (673–682)

Fundamentals of brazing for elevated-temperature service, M. D. Bellware (683–691)

How good is your welder power supply?, A. C. Johnson and F. E. Donathan (692–699)

Oxygen production, E. G. Hickling and A. E. Hittl (700–701)

Comparative properties of electrodes for arc welding austenitic manganese steels, W. L. Lutes and H. F. Reid, discussed by H. J. Chapin (702–705)

Composite arc-welded steel crankshaft devised for portable gangsaw, L. W. Johnson (706–707)

Prediction of weld heat affected zone microstructures from continuous-cooling transformation data, E. F. Nippes and E. C. Nelson (289s–294s)

Welding metallurgy of Cr–Mo–V steels for high-temperature steam-turbine components, R. J. Christoffel and others (295s–303s)

Stress relieving of stainless steels and the associated metallurgy, R. A. Huseby (314s–315s)

Improving the fatigue life of spot welds (315s–336s)

Role of phosphorus in austenitic manganese weld metal, W. T. Delong, W. L. Lutes and H. F. Reid (316s–319s)

Fabrication and service factors involved in failure of welded steam receivers, A. J. Babecki and P. P. Puzak (320s–325s)

Low crack sensitivity of steel joint by CO_2–O_2 arc welding, H. Sekiguchi and I. Masumoto (326s–336s)

Welding Journal (U.S.A.), 1958, vol. 37, August

Industrial applications of magnetic-flux gas-shielded arc welding, R. T. Telford and F. T. Stanchus (771–778)

Inert-gas-shielded arc welding of silicon and aluminium bronze, P. L. Hemmes (779–788)

Casting weldments in a petroleum refinery, J. Bland, C. B. Parrish and R. C. Wheeler (789–798)

Evaluation of spot welds made through primers and sealers, B. A. Schevo (799–802)

Diffusion bonding below 1000 deg F, J. T. Niemann, R. P. Sopher and P. J. Rieppel (337s–342s)

Control of melting rate and metal transfer in gas-shielded metal-arc welding. Part 1—Control of electrode melting rate, A. Lesnewich (343s–353s)

The control of porosity in high-nickel-alloy welds, G. R. Pease, R. E. Brien and P. E. Legrand (354s–360s)

The effect of welding speed on strength of 6061—T4 aluminium joints, E. L. Burch (361s–367s)

The influence of ferrite banding on the impact properties of mild steel, W. S. Owen, M. Cohen and B. L. Averbach (368s–374s)

Material rating based upon true stress-strain properties, J. Marin and M. G. Sharma (375s–378s)

Current welding research problems (379s–384s)

Welding Engineer (U.S.A.), 1958, vol. 43, mid-June

The meaning of weldability, T. B. Jefferson (5–20)

Classified product section (29–89)

Trade name index (90–111)

Manufacturers' names and addresses (112–121)

Welding Engineer (U.S.A.), 1958, vol. 43, July

TIG, gas, arc, et al: this is maintenance in a heavy industrial plant, F. Tancula (28–29)

Where maintenance welding often entails production problems, J. Fairlie (30–31)

Steel mills look to E3N12's for high strength in high heat (32–33)
Selection, welding of Cr–Mo alloy steel pipe, J. Bland (34–36)
Mig and Tig used to weld dynamite mixers (37)
Welding and stainless maintain water turbines, L. McWilliams (38–40)
Steel plant gondola cars fabricated by welding, R. Losee (41–42)
Arc versus gas welding on gray cast iron—which should it be?, J. C. Pellegrino (48)
Less cost and better appearances through welded bridges (50, 52)

Welding Engineer (U.S.A.), 1958, vol. 43, August

Filler metals for joining, O. T. Barnett (35–38)
The fundamentals of welding engineering, T. B. Jefferson (39–41)
Selection, welding of Cr–Mo alloy steel pipe (Part 2), J. Bland (42–46)
Introduction to welding: a course for cadets, P. G. Poetto (48–49)
Honeycomb gets new brazing method, R. B. Stanton (62–64)
Engineering data sheet No. 215: Comparison of spot welding tip areas when enlarged from original size (67)

Industry and Welding (U.S.A.), 1958, vol. 31, June

Six steps for dip brazing of magnesium assemblies, W. J. Graves (38–39)
60,000 lb of weld metal in world's largest crane, R. N. Williams (44–45)
Prestressed weldments are straight, C. Berka (47, 64)
New high-speed automatic welding process (51, 65)

Industry and Welding (U.S.A.), 1958, vol. 31, July

Furnace brazing: quality at low cost, H. M. Webber (34–37)
Production and engineering data: welding of cast stainless steel (38–39)
Special fixture speeds spot welding (43, 46)

Industry and Welding (U.S.A.), 1958, vol. 31, August

Should you set up for automatic welding?, G. L. Johnson (30–31)
Air cooled CO_2 welding (34–35)
Foundry welding and cutting, R. N. Williams (37)
Automatic metal spraying of precision parts (40–41)
Seven methods for joining stainless steels, P. Bowman (42–3, 63)
Thin-wall pressure vessel with 95% joint efficiency (46, 48)

Canadian Welder, 1958, vol. 49, May

World's biggest welding show (16–18)
The bridge on the river Grand (20–21)
Brazing filler metals for high temperature service (22–24)
L.A.'s "Dynaweld" welding process (26)

Canadian Welder, 1958, vol. 49, June

A report on the first National welding seminar (12–13)
Abstracts of papers presented at the first National welding seminar (14–20)
Operator participation in quality control, H. Howard (21)
Steel mine car bumpers redesigned for welding (22)

Canadian Welder, 1958, vol. 49, July

1958 Directory issue: Product listing section (14–42); **Manufacturer—Distributor reference** (American 43–50, Canadian 50–54, British 54)

Przeglad Spawalnictwa (Poland), 1958, vol. 10, June

The first welded bridge in Poland with built-up box section members, part 2 – design, Z. K. Lesniak
Welding procedure for alloy steel tubes, part 2, Z. Sepielak

Przeglad Spawalnictwa (Poland), 1958, vol. 10, July

The influence of some weld defects on the strength of statically loaded welded joints, T. Nawrot
Weld cracks in a railway bridge, A. Fabiszewski
Alloy steel tubes welding procedure, part 3, Z. Sepielak

Przeglad Spawalnictwa (Poland), 1958, vol. 10, August

Powder cutting, part 1, W. Szydlik
Repair of Kaplan water turbines by welding, without dismantling, L. Dreher and S. Klosowksi
Repair of low-alloy steel castings in turbine construction, R. Pasierb and M. Dymczak

Journal of the Japan Welding Society, 1958, vol. 27, June (Abstracts in English)

Fundamental study on buckling of thin steel plate due to bead welding, M. Watanabe and K. Satoh (13–20)
Severe wear of pearlite steel weld metal, S. Ito, K. Honda and K. Ishiyama (21–26)
A fundamental experiment of arc welding process with superposed high-frequency current, I. Ukita and S. Nobuhara (27–30)
On the CO_2—Sekiguchi's filler wire arc-welding process (Report 7), H. Sekiguchi and I. Masumoto (31–40)
Continuous cooling transformation diagrams of various steels submitted to welding (Report 1), H. Sekiguchi and M. Inagaki (40–46)
Resistance welding of commercially pure Ti and its fatigue strength, T. Nakamura and E. Maeda (47–54)

Journal of the Japan Welding Society, 1958, vol. 27, July (Abstracts in English)

A study of inert-gas metal-arc welding by high-speed motion pictures (Report 1), I. Ukita and S. Nobuhara (13–18)
Considerations on the strength of inclined fillet welds, S. Ohta (19–25)
On the CO_2—Sekiguchi's filler wire arc welding process (Report 8), H. Sekiguchi and I. Masumoto (26–31)
Continuous cooling transformation diagrams of various steels submitted to welding (Report 2), H. Sekiguchi and M. Inagaki (32–38)
Data used for stationary large-capacity acetylene apparatus (Report 1), I. Ueda, Y. Imai and Z. Murakami (39–43)
Effect of phosphorus in parent metal on the weldability of austenitic manganese steel, S. Ando, N. Kimata and H. Doizaki (44–51)
Strength of brazed joints, T. Yoshida and N. Nomura (52–55)

Welding News (Australia), 1958, Folio 94, April

Story in steel (2–4)
Oxy-cutting knows no limits (5–9)
Iron powder electrodes (10)
Oxygen: the flame of a thousand uses (15–22)

Welding News (Holland), 1958, No. 89, March

Philips C 6, P. C. van der Willigen (2–5)
The Indoor-skating rink at Boulogne-Billancourt (6–10)
The welding of long studs, J. A. N. Clevers (11)
Stud-welding: the fixing of electricity supply tubing and wiring systems, J. A. N. Clevers (12)

Zvaranie (Czechoslovakia), 1958, vol. 7, July

Special kinds of corrosion connected with the welding of Anticorro and heat-resisting steels, J. Nemec (193–196)
Possibilities of welding circular sectors into molten slag, J. Zeke (196–200)
Principles, advantages and limits of dielectric welding of softened thermoplastic foils, L. Dvorak (200–206)
Improved striking of the arc in automatic welding, V. Sulc (207–208)
Contribution to the rectification of distortion by heating with different thermal sources, J. Matas (219–220)
Gas welding in repairs of cast iron machine parts, V. Mutnansky (212–218)

Zvaranie (Czechoslovakia), 1958, vol. 7, August

Material for the jaws of resistance flash welders, R. Rohan (227–233)

The possibilities of surfacing trench-holding knives of coal cutting machines, J. Kolar (234–237)

A new type of suspended spot welder, L. Pliva (238–244)

A contribution to brazing with the brazing solder Cu–P, P. Slysko (245–248)

Safety in welding, P. Hrbal (249–251)

Technology of repairs of slag inclusions in railway wheelbands by welding, L. Muncner (251–252)

Australian Welding Journal, 1958, vol. 1, May

Some techniques in the welding of gasholders, Mrs. N. G. Mather (7–11)

Arc welding electrodes: large or small? (13–15)

Australian Welding Journal, 1958, vol. 1, June

Mammoth girders and columns for powerhouse fabricated by Twinarc (12–13)

Four unusual all-welded Vierendeel spans (13)

Preparation and fit up for welded joints, B. Osman (15–17)

Low temperature hard overlaying in railway maintenance (18)

Complete site-welded construction of a 14-storey building, K. H. Horwood (25–26)

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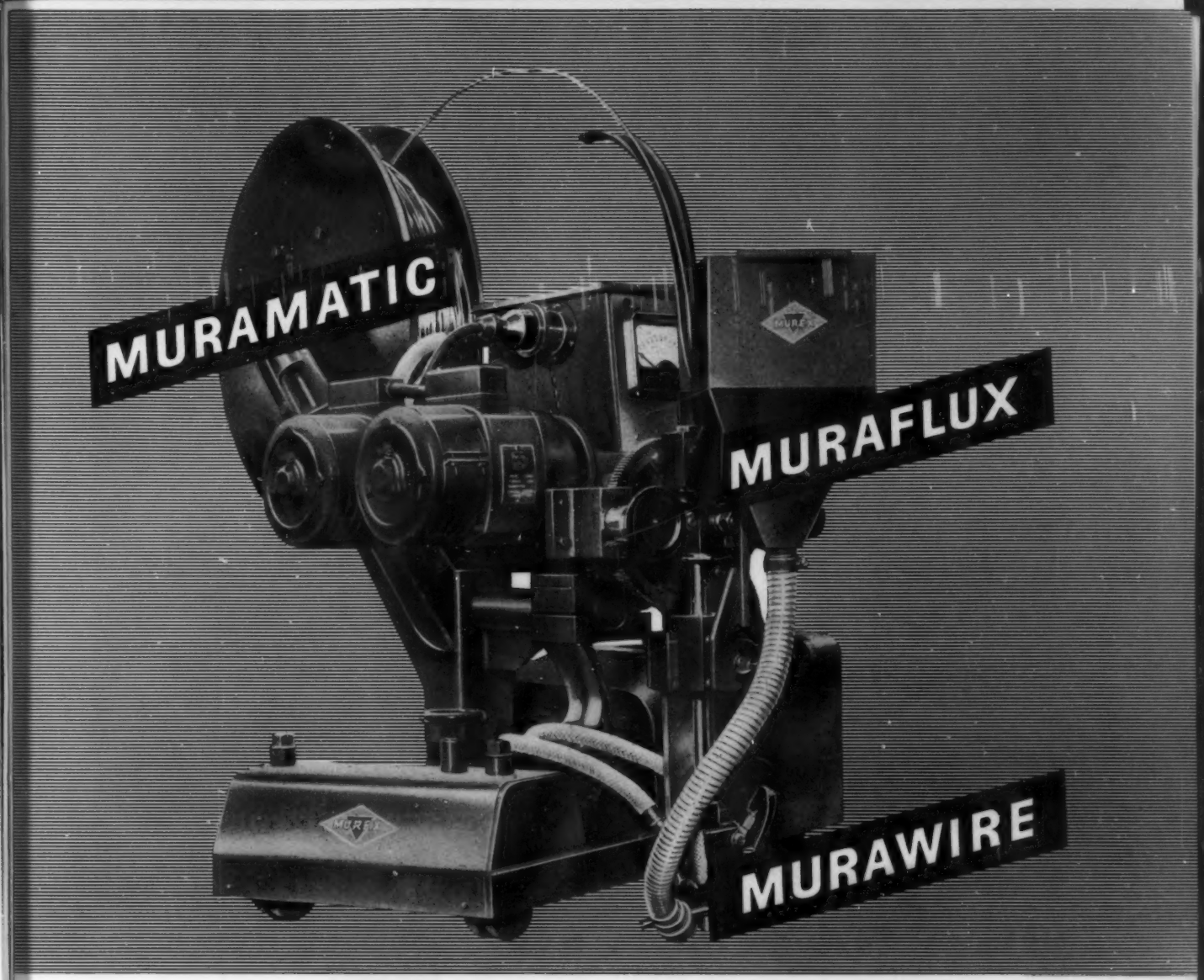

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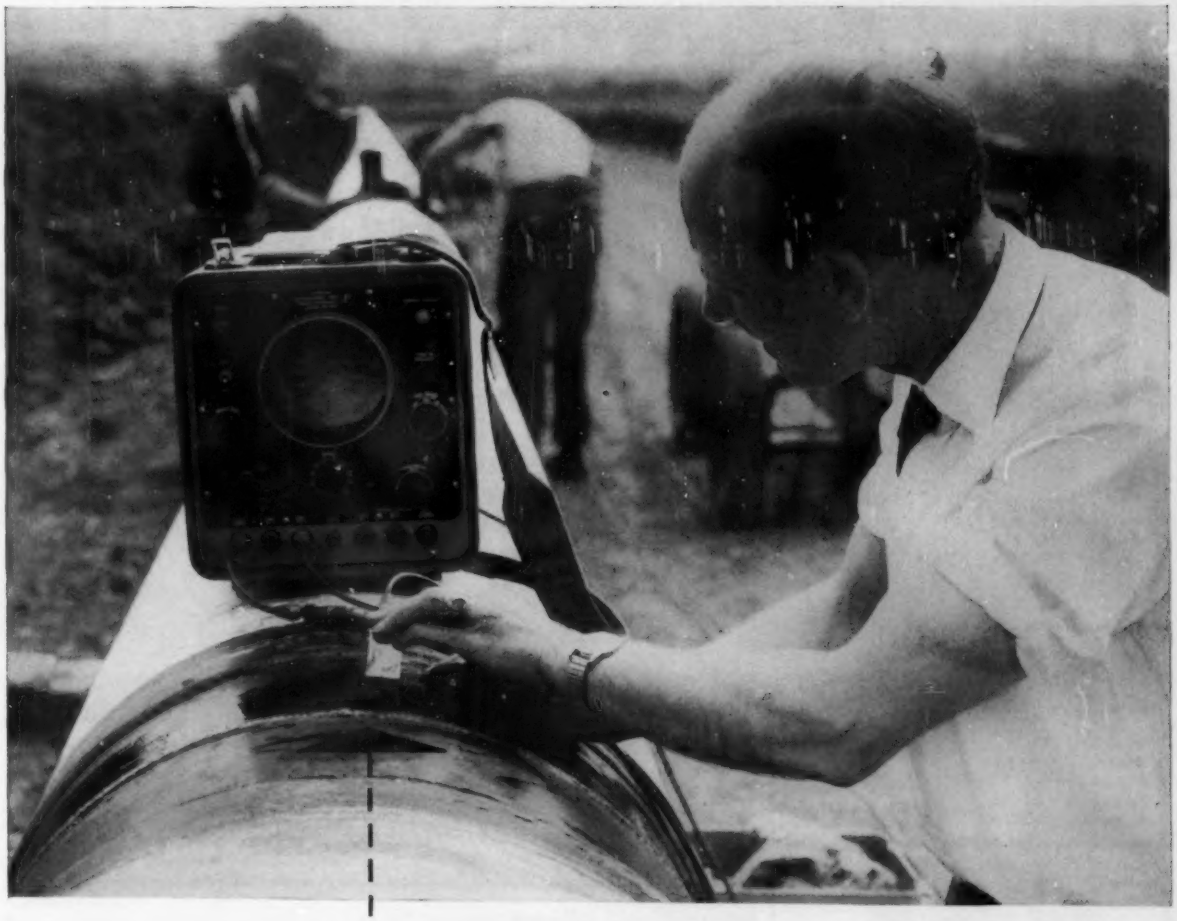

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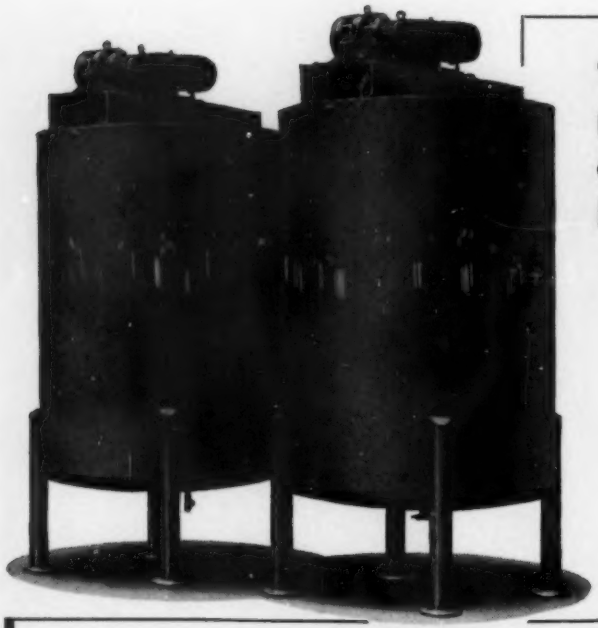

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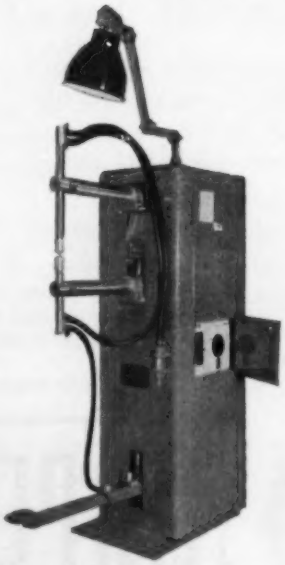

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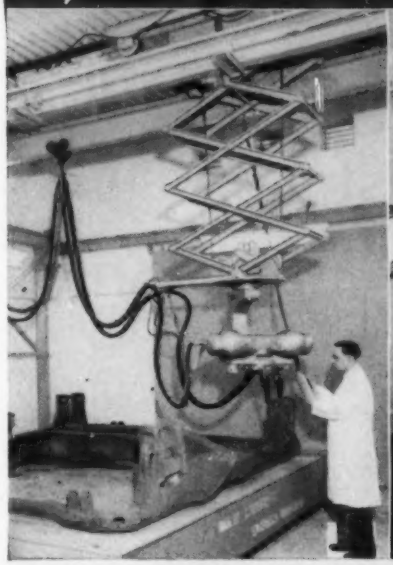

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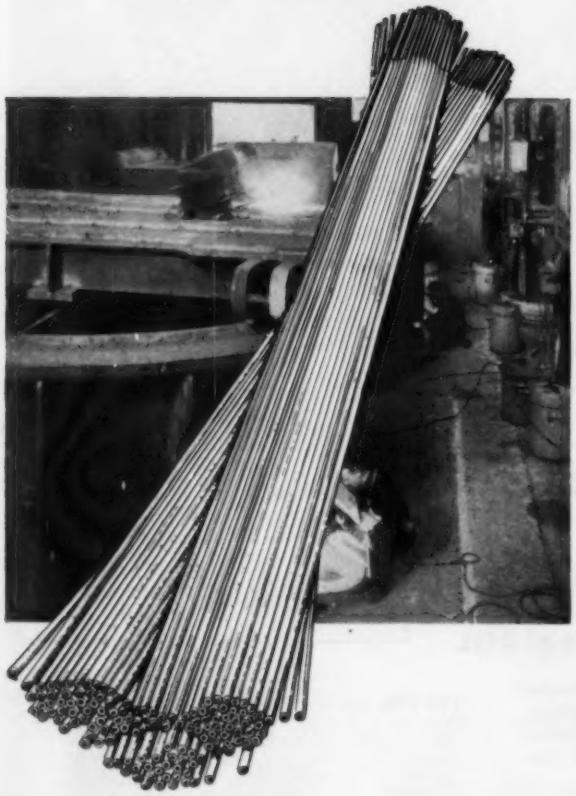

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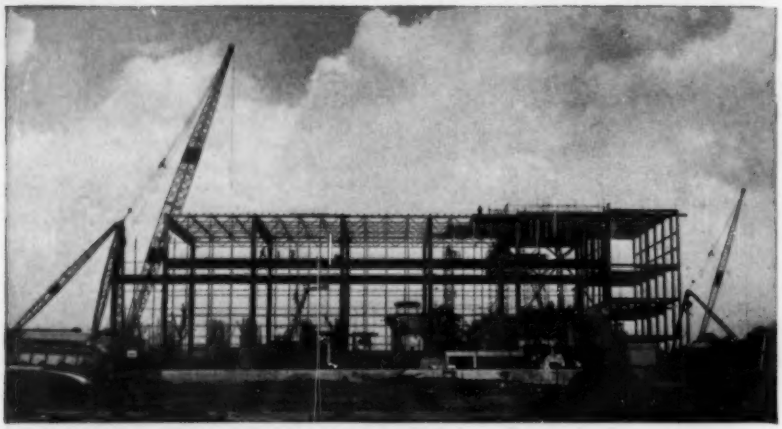

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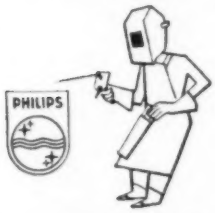

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